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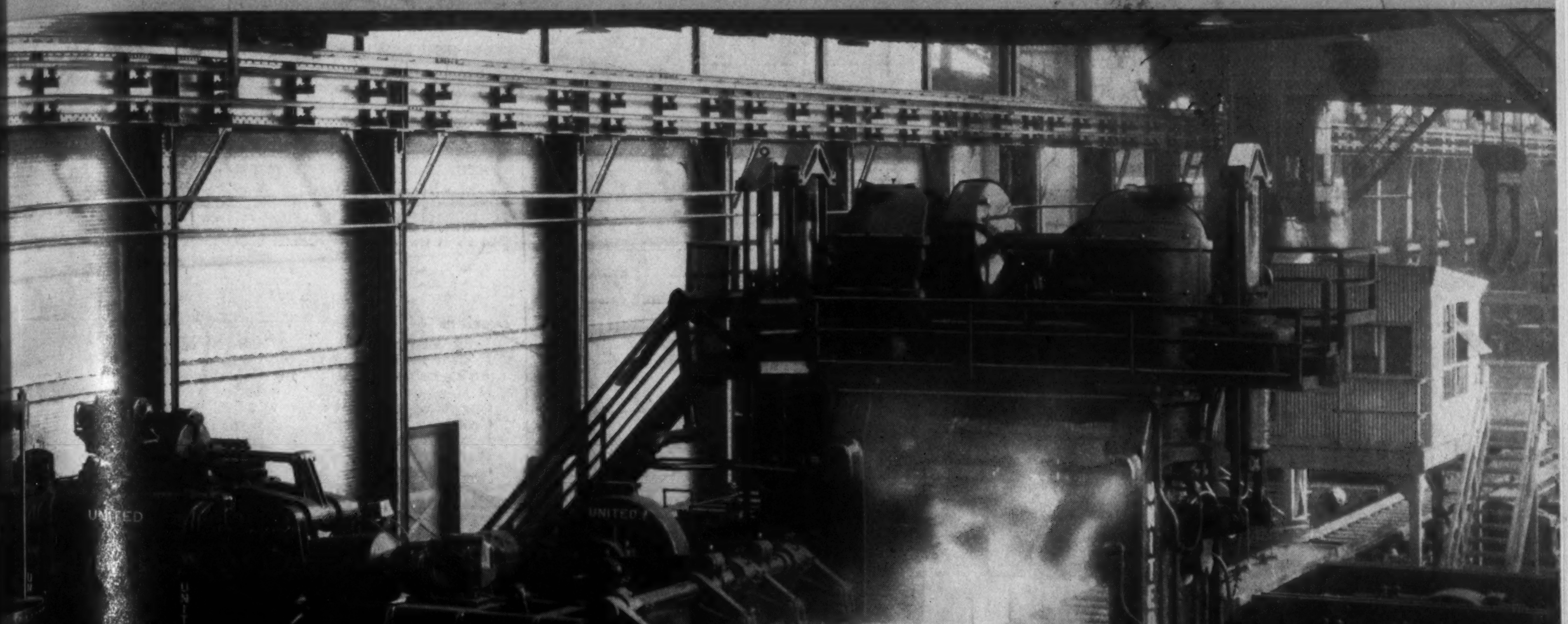
The Magazine of Metallurgical Engineering

INCLUDING

CURRENT METALLURGICAL ABSTRACTS

4 1932



40″ 2-High Reversing Blooming Mill of A. M. Byers Co., Ambridge, Pa. (*Courtesy United Engineering & Foundry Co.*)

VOLUME 3

$3.00 a Year

OCTOBER
1932

NUMBER 10

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NICKEL BRONZES

The close metallurgical similarity between nickel and copper, shown in one respect by the unlimited degree to which these metals alloy with each other, is also reflected in the amazingly large number of nickel alloys having a copper base which have at one time or another found favor with engineers and with artisans.

In general, nickel dissolves in, toughens and hardens the alpha phase of copper alloys. The ease with which it is taken up by the molten alloys at temperatures far below the fusion point of nickel; its complete retention without oxidation losses; its freedom from tendencies to segregate or liquate, leave few problems for the foundry engineer.

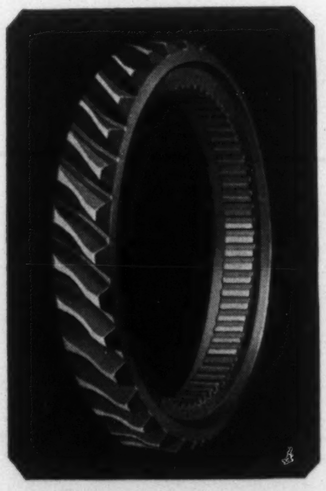
Those bronzes containing 10% to 30% nickel provide properties closely associated with nickel, as great toughness, high strength at atmospheric and elevated temperatures, corrosion resistance and whiteness of color.

With nickel contents of 3 to 10% bronzes having superior high temperature properties as required in steam engineering are secured, and with appropriately controlled tin contents, a series of bronzes of great hardness have been developed.

Small nickel contents, of ½ to 3%, are of value chiefly in modifying the mechanical and casting properties, and are usually employed as additions to the common types of bronze foundry mixture.

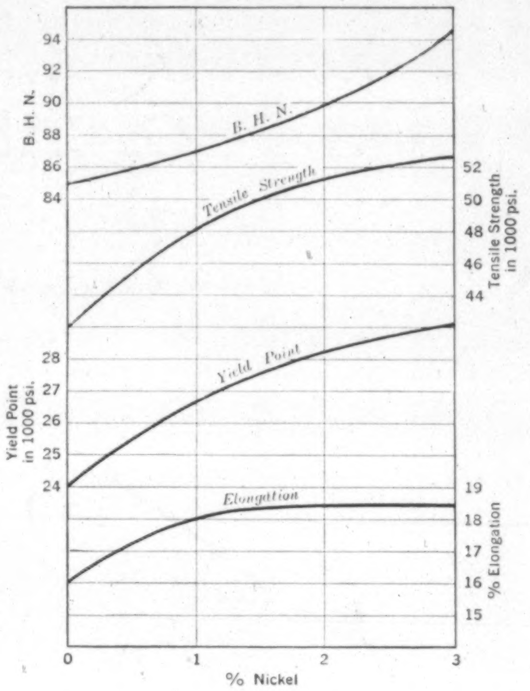
Foundry Bronzes

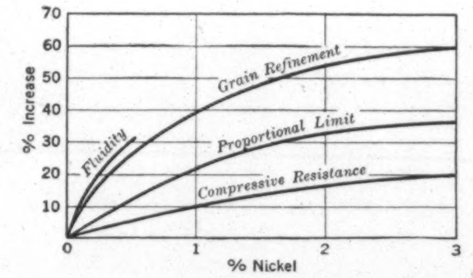
While an almost endless variety of combinations of tin, lead, zinc and other metals with copper comprises the bronze foundry alloys, most needs of users of bronze castings are met by less than a dozen types of mixture. These alloys—red and yellow brasses, gun metal, leaded bearing bronzes, tin bronzes, etc.—respond in similar directions to modifying nickel additions, the most effective content of which depends both upon the detailed composition of the bronze as well as the use to which it is to be put.



Nickel Bronze Automotive Worm Gear Containing Approximately 1% Nickel

Typical of the modification in properties are the data shown in the chart below for a bronze containing 11.1% tin and .03% phosphorus, a type frequently used in the manufacture of gears.





Improvements Effected

Nickel in amounts of ½ to 2% favorably modifies:

(1) Elastic properties
(2) Tensile strength
(3) Compressive strength
(4) Hardness
(5) Toughness
(6) Impact resistance
(7) Grain size
(8) Fluidity
(9) Density
(10) Lead distribution

Mechanical Properties

The strengthening action of nickel is experienced most effectively with slight degrees of deformation, evidenced by a rise in proportional limit and yield point which is relatively greater than the corresponding change in tensile strength or hardness. A 25% rise in proportional limit is not unusual, and the importance of this stiffening characteristic in connection with castings which may fail by distortion rather than by fracture is of course evident. The rise in tensile strength, compression strength and hardness is generally obtained without impairing ductility, and frequently with enhancement of toughness. Nickel increases the resistance of many bronzes to sudden shock, developing a 25-40% increase in the Izod impact test values.

Casting Properties

More elusive but equally important are the casting properties; although nickel increases slightly the melting point of bronzes and might be expected to diminish the ease of filling all parts of the mold, the reverse is true. A number of common mixtures including tin bronze and red brasses, show increased lengths of run in the spiral fluidity test as great as 40% with as little as ½% nickel addition. Fewer misruns, widened casting range and lower pouring temperatures result from this characteristic. A similar content of nickel reduces the tendency toward formation of intercrystalline shrinkage pores, an action felt as increased density, improved fracture and—with hydraulic mixtures—increase in pressure tightness and reduction in number of "leakers". The grain refinement secured by similar small percentages of nickel in sand castings within the normal pouring temperature range is from 40 to 50 per cent, as indicated by the decrease in average grain diameter. In highly leaded bronzes, nickel minimizes lead segregation and accomplishes a finer dispersion of this important constituent.

Uses

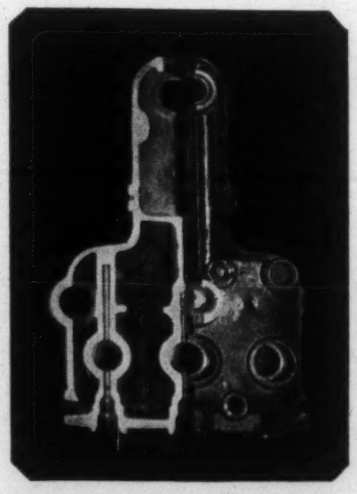
Bearings (½-2%)
- Engine bearings
- Rolling mill bearings
- Automotive bearings
- High speed bearings
- Car journal bearings
- Crankshaft bearings
- Locomotive brasses
- Bushings
- Bridge bearings
- Turntable bearings
- Shaft bearings
  - Stationary
  - Marine
  - Aero

Valves and Fittings (½-3%)
- Globe valves
- Gate valves
- Sea valves
- Valve trim
- Steam valves
- High pressure valves
- Superheated steam valves
- Steam turbine valves and seats
- Air, gas and water fittings
- Slide valves
- Hydraulic castings

Miscellaneous
- Gear blanks
- Worm wheels
- Lubricator parts
- Injectors
- Paper mill rolls
- Paper mill rotary screens
- Gun recoil sleeves

Sources of Supply

Nickel bronze castings are regularly produced by leading non-ferrous foundries who are ready to assist in the selection of compositions best suited to meet particular requirements. If the improved properties of nickel bronzes are of interest to you we shall be glad to put you in touch with established sources of supply.



Nickel Bronze Locomotive Lubricator Casting Containing about 1% Nickel

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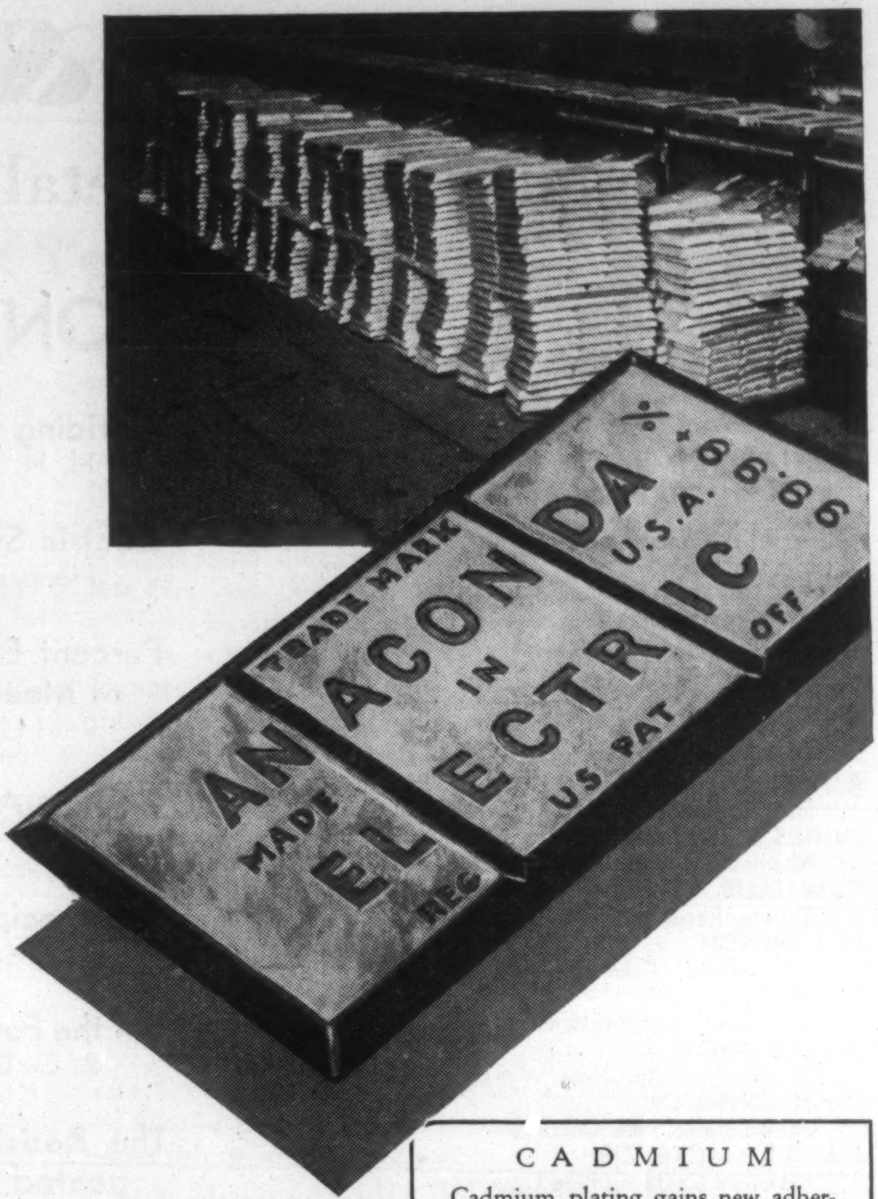

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PUBLISHED MONTHLY BY THE CHEMICAL CATALOG COMPANY, INC.
Ralph Reinhold, Pres.-Treas.
Philip H. Hubbard, Vice-Pres.
H. Burton Lowe, Vice-Pres.
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Annual Subscription: U. S., Possessions, Mexico and Cuba, $3.00; All Other Countries, $4.00. All communications relating to subscriptions or back issues should be addressed to METALS & ALLOYS, 3619 Forbes Street, Pittsburgh, Pa.
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Entered as second-class matter March 14, 1932, at the post office at Pittsburgh, Pa., under the Act of March 3, 1879.

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EDITORIAL COMMENT

Get-Rich-Quick Testing Methods

MAN IS a lazy animal. He is prone to take the road that seems to offer the shortest path even though it may not bring him out at his destination. The testing engineer is beset with demands for short-cut tests on the properties of metals and alloys on which it would take many years of service to establish the facts. He is all too prone to yield to these demands, especially in relation to corrosion, wear, endurance under repeated stress, and creep at high temperatures.

If someone will only say that an accelerated test method appears to be reliable, other people will rush to use the short-cut method, usually without first finding out its range of applicability. Then, when a lot of short-cut data have been accumulated, more conscientious workers take the joy out of life by showing that the short-cut data are unreliable and often dangerously misleading. On the whole, the short-cut methods are, like other get-rich-quick schemes, likely to be profitless in the end.

We have seen, in the field of corrosion-testing, the man who wants to establish the superiority of his particular sheet for atmospheric corrosion, attempt to do so by sticking the samples in acid. This situation got so bad that the A.S.T.M. had to run a lot of expensive and technically unnecessary acid tests to give a scientific background to its official statement[1] disclaiming "any recommendation or endorsement of the acid test as a measure of natural corrosion."

Wear-testing is in a chaotic state, but one can be quite certain that much of the wear-test data in the literature, carried out by rapid methods, are quite unreliable in respect to actual service, since the conditions do not at all duplicate those of service, nor produce the same type of wear.

Fatigue testing at very high stresses has its place to establish the ability of a material to stand unusual punishment, but has little relation to performance under service stresses. The work of Moore and many others has shown how to make reliable endurance tests and fortunately, for most classes of steels, 5 or 10 million cycles will give useful data. It is something of a shock, since the facts about endurance testing are so well known, to find in a fairly recent book,[2] the statement, made in italics, that "it is only necessary to make a Brinell hardness test on a specimen of steel in order to determine its endurance limit." Rosenhain[3] points out, however, that while it is "a tempting way out of the difficulty to rely on the ratio of endurance limit to tensile strength," (or Brinell hardness), this ratio is likely to be seriously affected by the presence of defects, and endurance limits calculated from it would be wrong, since the presence of "stress-raisers" is not readily indicated by anything but a true endurance test.

Elsewhere in this issue[4] we discuss the uselessness of accelerated endurance "tests" on copper-base alloys, and the necessity for running endurance tests to many millions of cycles.

The accurate determination of the high-temperature properties of metals, especially their resistance to creep, is so tedious and expensive that various efforts have been made to calculate the probable behavior during years of service from observations in, say, the first 50 hours or less of a test. Various methods have been put forth, some of them having been slightly modified later by their proposers. Quite a nice correlation is shown in some instances between the results of short-cut methods and those of true creep tests and the non-critical reader is likely to get all "jazzed up" about the possibility of avoiding a huge "waste" of time in running real 1,000 or 2,000 hour creep tests.

However, Pomp,[5] an advocate of a short-cut creep method, has recently reported results on the accelerated method, as compared with a true creep test, on a heat-treated Ni-Cr steel which show that the short-cut method leads to entirely erroneous conclusions, and to conclusions in error on the unsafe side.

This, coupled with plenty of other evidence to the same effect, confirms our belief that all claims for reliability of short-cut creep tests should be taken with a very considerable grain of salt.

The trouble with the proponents of short-time tests is that so many of them go off half-cocked. They take a sample of mild steel, make a short-cut creep or fatigue test of it, find rather good agreement with the standard test, and jump to the conclusion that the short-cut method applies to everything. It would be fine if rapid methods could be found, and the search for them should not be discouraged, but until they have been proven out, they should always be taken as only on trial, for purposes of "testing the test," and never as reliable data for engineering design.

Perhaps the depression has made the scientist less gullible and less likely to give credence to get-rich-quick schemes, whether in the realm of finance or of testing. We hope so, anyhow.—H. W. GILLETT

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[1] Report of Committee A-5 on Corrosion of Iron and Steel. *Proceedings of American Society for Testing Materials,* Vol. 29, Part 1, 1929, page 150, Vol. 31, Part 1, 1931, pages 171-180.

[2] A. W. Judge, Engineering Materials, Vol. III, Theory and Testing of Materials, 1930, page 183.

[3] W. Rosenhain, Value, Scope and Interchangeability of Mechanical Tests. *Metal Progress,* June, 1932, page 58.

[4] Page 236.

[5] A. Pomp & W. Höger, Dauerfestigkeitsuntersuchungen an Kohlenstoff-und Niedriglegierten Stählen nach dem Abkürzverfahren. *Mitteilungen Kaiser-Wilhelm Institut für Eisenforschung,* Düsseldorf, Vol. 14, No. 4, 1932, pages 37-57.

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RICHARD RIMBACH, *Editor*

Nitriding Plant and Technique

By H. H. ASHDOWN*

DURING the past few months a number of excellent articles have been published on the subject of nitriding. The principles of the process have been well described as also have the materials most receptive and responsive to this process.

It is not the intention in this article to recapitulate the information already referred to, but to discuss the equipment, apparatus and technique requisite for obtaining on an economical production basis the highest efficiency from the process.

The process as at present operating commercially is one requiring a comparatively long period dependent on the depth and hardness of case required. It is practically automatic throughout this period requiring little if any personal attention, apart from an occasional survey of the controlling instruments. It has been made to appear a simple process and it is a simple process once one has become conversant with its limitations and the apparatus essential. It, like many other processes, is simplicity itself when all goes well, but there are many simple factors involved, any one of which, failing to function, may lead to dire disappointment. The Westinghouse Electric & Manufacturing Co. at the inception of this process, foresaw the wide scope of the application for nitrided materials and at the same time realized the need of perfect control which demanded the best and most reliable equipment, particularly the need of a perfectly balanced heat control.

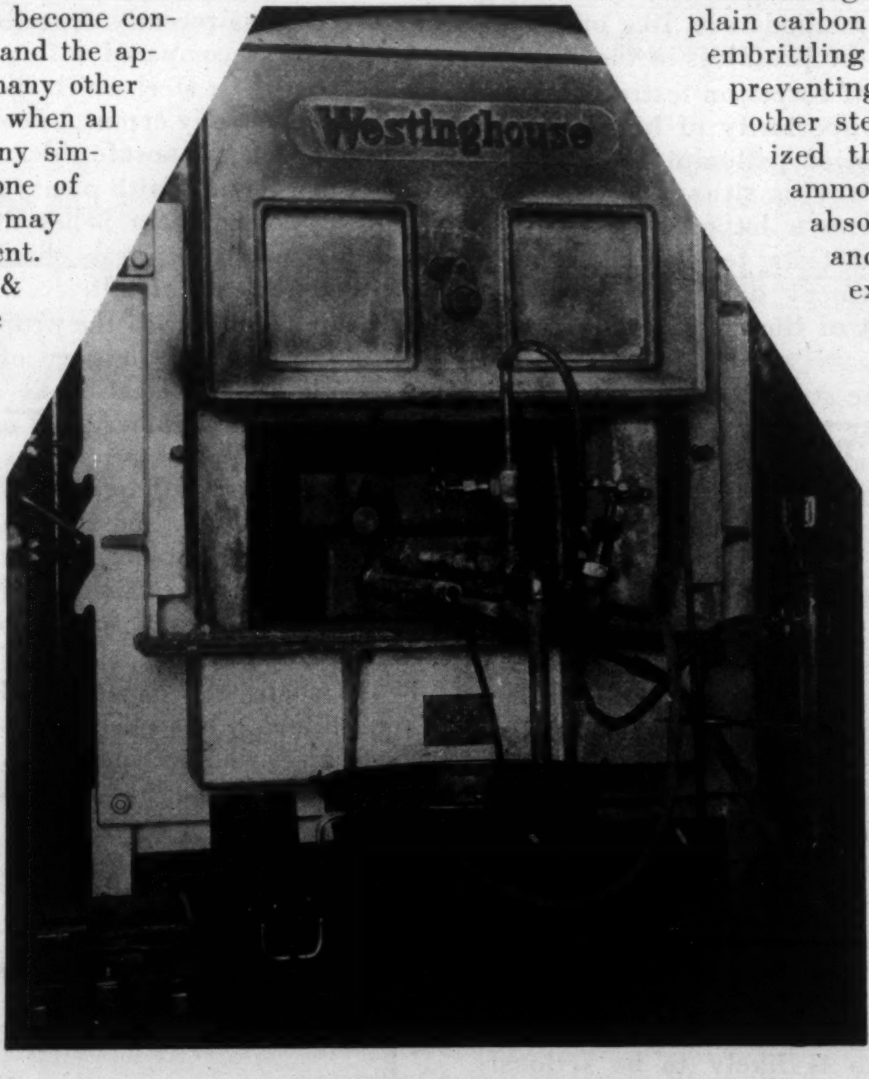

Fig. 1.

Being manufacturers of all details for electrical apparatus and equipment, ranging from the domestic house switch up to the largest power generator units, in many of which nitrided parts are used and also manufacturers of electrically heated furnaces of equally wide range, both fields have been fully explored. As a result of costly experiments and practical experience of several years, such technique has been developed and apparatus produced that a very high standard of quality and regularity of production has been attained.

It is well known that this process is the result of the research by Fry of Essen, but it has and still is receiving considerable attention in our research laboratories and production departments. However in transferring a process from the research laboratory onto a commercial production basis, many new factors present themselves and require experience and a carefully developed technique. In Fig. 1 is shown our works experimental container measuring 12″×10″×18″ in which development work is carried out and in which also small urgent production orders are executed. This further lends itself to the treatment of small quantities of detail parts for either short or long cycle periods and to various predetermined nitriding conditions.

Early experimental and production work was conducted in containers formed of commercial steel plate and the manifolds for ammonia distribution, exit gases, and the sheaths for thermocouples also were of like material. It soon was realized that although little hardening effect was produced on this plain carbon steel a comparatively rapid embrittling effect resulted throughout, preventing adjustments to these and other steel fittings. It also was realized that large quantities of the ammonia gas were dissociated and absorbed by the container walls and incidental equipment at the expense of a more satisfactory product. Of all the materials carefully explored for built up containers few have offered a satisfactory solution as a commercial proposition. The vitreous enameled surface due to repeated changes in temperature, crack and flake at the corners. Monel metal unless of low manganese content shells readily and an electro deposited nickel surface is apt to be spongy and somewhat pervious to the activating gas. Straight nickel plate has presented many problems in obtaining satisfactory welded joints to withstand the service conditions of the container.

Although certain types of Monel metal and some of the nickel chromium alloys offer some promise as suitable materials for the containers, pure nickel fittings, racks, trays, etc., which have been subject to the activating gases for more than 1000 hours still retain their original malleability and appear to be unaffected by the activating gases. From past experience it would appear that a satisfactory and economical solution for container construction is to use nickel faced rolled steel plate with the nickel ply face of 15 to 20% of the plate thickness. After forming to shape, the outer joints of the carbon steel plate are readily welded which offer rigidity to the subsequently welded portion of the joints of the nickel face which together are impervious to gas penetration.

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*Metallurgist, Westinghouse Elec. & Mfg. Co., East Pittsburgh.

Both experimentally and together with production work a large variety of materials was submitted to the nitriding cycle and in addition, the method of nitriding was rigorously studied with a view to determining the best and most efficient conditions in order to obtain the best product, in the direction of maximum hardness, toughness and case depth. Base materials which retained the best physical properties of the core together with the case properties already enumerated and at the same time those which responded best under the most economical conditions of operation were studied. It was found that material giving an approximate analysis of 0.30 C, 0.25 Si, 0.80 Mn, 0.75 Mo, 1.0 Al nitrided in a complete cycle of 25 hours at a maximum temperature of 580°C. and produced a non-brittle case of approximately 0.02″ deep with a Vickers Brinell numeral around 950 under a 50 kg. load, while other nitriding materials available required about a 40 hour cycle to acquire similar results. Of comparatively recent date other alloy steels have been so modified in their composition as to bring them more nearly within the shorter nitriding cycle.

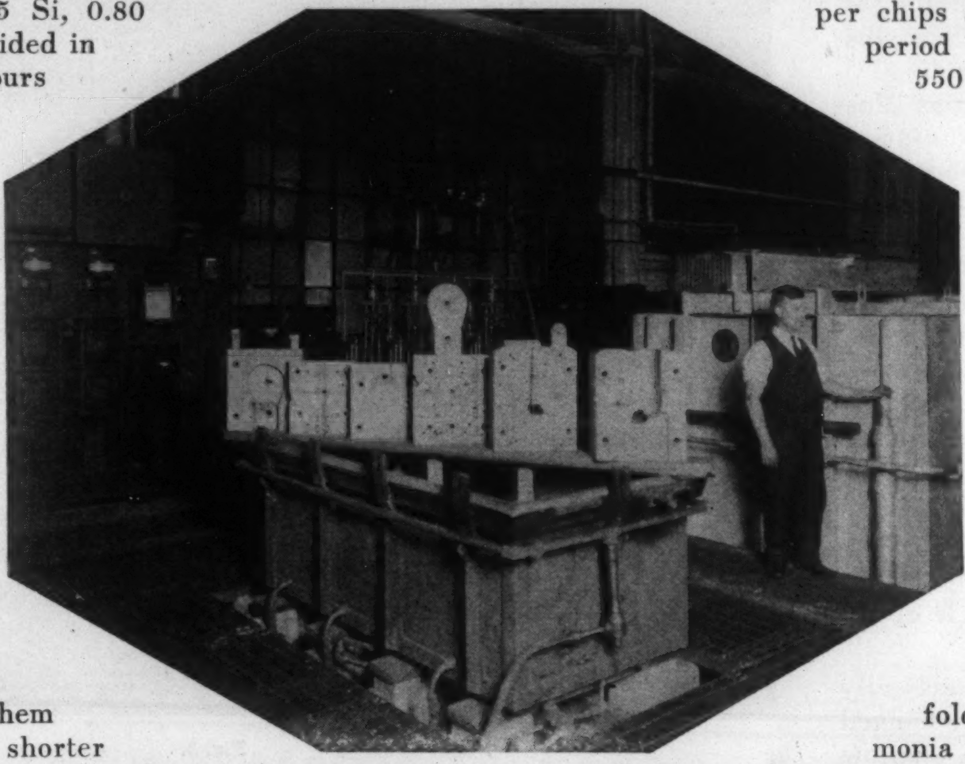

Fig. 2.

Due to the extremely hard and high wear resistant properties of the case, for all practical purposes a case depth of from .015″ to .02″ is fully adequate. Penetration beyond this depth is of little practical value and as the rate of further penetration is relatively slow it adds considerably to the production costs.

Of all the steels now available for this purpose, although excellent results may be obtained from them in the heat-treated condition, superior results in both cases and core are obtained subsequent to a quench and tempering treatment. It should be noted that the tempering heat must slightly exceed that of the maximum nitriding heat when definite physical properties of the core will be known and all such treated parts will be free of any distortion subsequent to nitriding. It is however imperative that a machining operation should follow any heat-treatment in order to insure the complete removal of the decarbonized surface, as the presence of such is fatal to good nitriding results.

It also is equally important that all traces of grease or foreign matter be removed from parts to be nitrided which may be done by washing in gasoline or hot alkali solution and a final wash in hot water, particular attention in this respect being given to all recessed or hollow parts.

It has long been known that a notched case hardened specimen, however tough the core may be, when subject to a sharp blow will snap off with a clean break. This applies equally to nitrided steels and for that reason any detail parts, particularly those subject to shock stresses, should be free from perceptible tool marks. Further it is advisable wherever possible to introduce fillets and avoid sharp angles.

It is generally known that a "growth" of the material takes place during nitriding and for the average depth of case it may be assumed to be one fifteenth that of the depth of the case for solid parts, for example a case depth of .02″ will give approximately a growth of .0016″, although for rings and hollow parts the growth internally is somewhat less. Attention is drawn to this phase as the growth on sharp angles and pointed parts is more pronounced and the case appears to be more brittle. Where sharp corners, etc., are essential, this brittleness may be relieved by burying these parts in spent carburizing material or fine steel or copper chips and heating for a short period to a temperature of 550° C. This treatment effects a slight inward diffusion of the nitrides, resulting in a toughening of the case generally and has little if any effect on its maximum hardness.

Fig. 2 illustrates a container measuring 28″×34″×72″ typical also of a larger container measuring 30″×30″×108″ which are representative of built up steel nickel clad interiors. The manifolds discharging the ammonia gas into the container are of pure nickel as are also the thermocouple sheaths and the exit manifold shown at the top of the container. The base manifolds are finely perforated to diffuse uniformly the ammonia throughout the charge and are so arranged to effect dissociation at the lowest possible position in the container.

In some types of nitriding containers it has been found expedient to insert a circulating fan to create turbulence of the activating gas, this however has a dilution effect as both hydrogen and inert nitrogen are unnecessarily retained, which may slow down the process. By the Westinghouse method of electrically heating the container with heating elements which are both under the base and surround the walls a uniform and consistent temperature is maintained. The heating from below insures immediate dissociation of the incoming ammonia, releasing the active atomic nitrogen immediately under the work and due to the normal upward flow the decomposed diluent gases are swept through the exit manifold, leaving the more concentrated active gas around the work.

Apart from proof of the uniformity of the work as taken from the container, check tests are made by suspending strings of buttons of the same material 4″ apart from top to bottom of the container and in various positions and from these confirmatory evidence is obtained of the uniformity of products within 3″ of the bottom or 4″ of the top, although as a further insurance factor a wider margin may be given to the work.

One of the great advantages of this type of stationary container is that the work can be placed in position on nickel frames, racks, or trays according to type, and once so placed, is not disturbed until the nitriding cycle has been completed and the work ready for removal. The value of this will be better appreciated when considering the care necessary in handling airplane or Diesel engine crankshafts. These are ground to size on the journals, etc., and packed in position which in the trolley type furnace may by a slight jar, be readily disarranged, thus causing distortion during nitriding.

Again the bell furnace as illustrated in Fig. 2 is of a continuous type inasmuch as the containers are in duplicate but on separate bases. The walls of the bell furnace carry the heating elements and immediately the nitriding cycle has been completed in one container the electric cables are disconnected, the bell raised and lowered over the companion container, all the residual heat in the bell being transmitted to this second container. The necessary electrical connections are made externally and the same cycle of operations follows:

The containers are hermetically sealed by inserting the flanged edge of the cover into a shallow trough containing molten lead or other low temperature melting alloy, which surrounds the extreme top of the container. After the temperature of the container has fallen to approximately 150°C., the metal may be pulled out from the seal or remelted for which suitable provision is made and run off through an aperture provided for this purpose. The finished nitrided parts present a silver gray color but large sections even at this temperature when exposed to the air will sometimes acquire a slight discoloration ranging from straw to blue corresponding with the well known range of steel draw heat colors. This discoloration however, apart from appearance, is of no consequence as it has no effect on either the hardness of the case or the material as a whole. Reference has already been made to the nitriding cycle and it has been found that for loads of around 2000 lbs. of miscellaneous parts it requires approximately five hours to raise the temperature to 550°C. which temperature is maintained for about 8 hours and then gradually raised to 580°C. and finally maintained at this heat for about 12 hours. By this treatment a depth of case of approximately 0.02″ is obtained with a Vickers Brinell hardness around 950. The dissociation at these heats being respectively approximately 25 and 50 cc. of activated ammonia. As previously stated, each installation will present its own particular problems, a solution for which can only be found on the spot and once surmounted the process has few equals in simplicity.

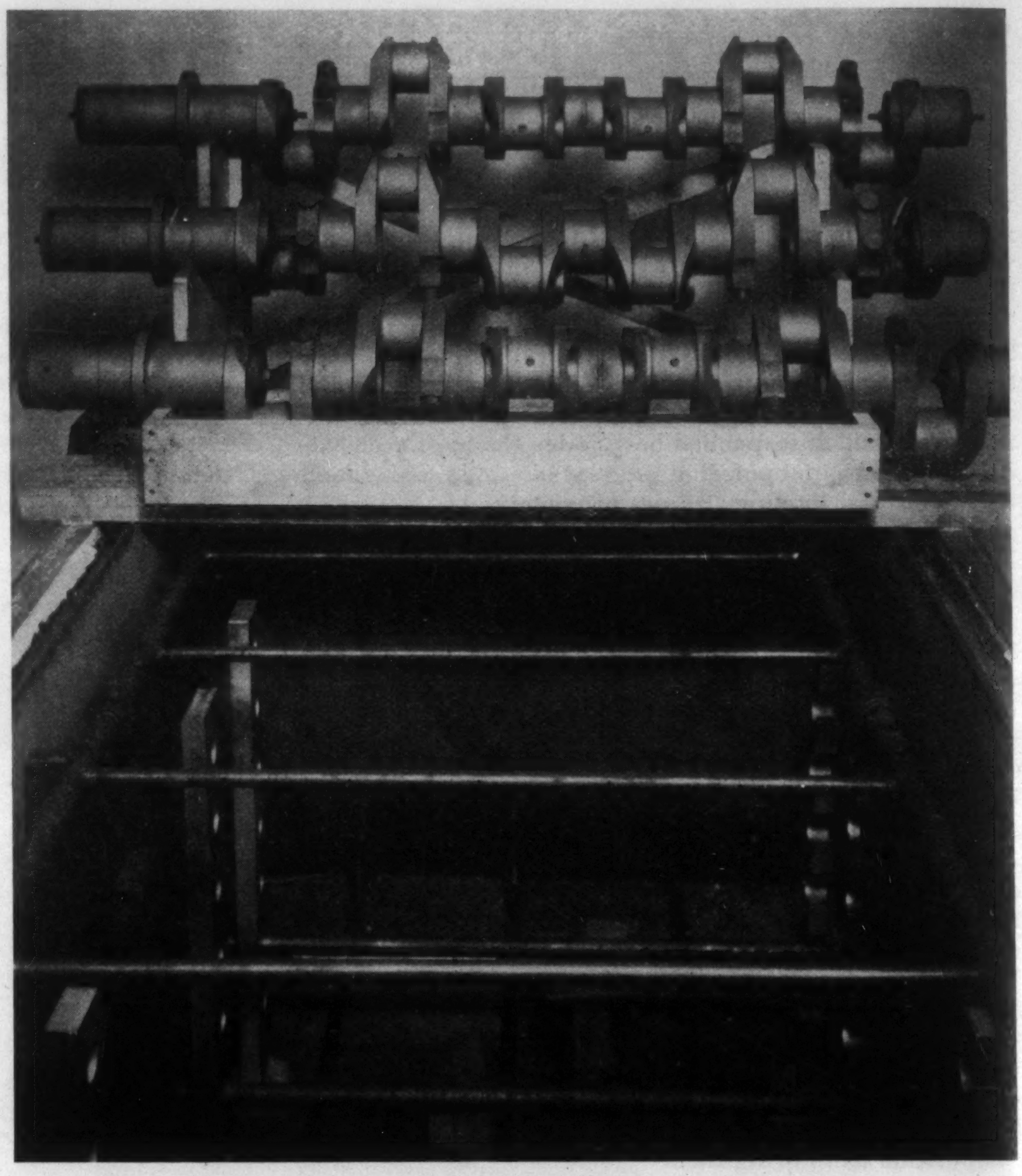

Fig. 3.

One of the great advantages of this process over any of the carburizing (case hardening) methods is that on completion of the cycle, the product is finished ready for assembly. No quenching operation is necessary, therefore for intricate parts such as die casting dies, gears, or those of marked changes in section, there is no risk whatsoever of the slightest distortion, hardening cracks, or residual stresses.

Nitrided surfaces are resistant to atmospheric corrosion and also to weak acids and alkalies which fact makes the process valuable in a wide field of application.

In the background of Fig. 2 will be seen all the branch piping and reducing valves conveying the ammonia gas to the 4 separate base units in this installation. Each of the 4 branch pipes is again subdivided into 3 and in each case these 3 pipes are grounded and pass along a culvert to the base of each container where they join up to the bottom manifolds distributing the gas uniformly within the container. The square boxes near the base of the piping are water containers through which the exit gases pass and serve both as an indication of the rate of flow of gas and also absorb much of the free undissociated ammonia. It is at this point also that connections are made for periodic checking of the dissociation. Figs. 3 and 4 give typical examples of charges as removed from the containers after nitriding. In Fig. 3 may be seen the top manifold in position in the container.

So important has nitriding become to the Westinghouse Company that it was found necessary to make such

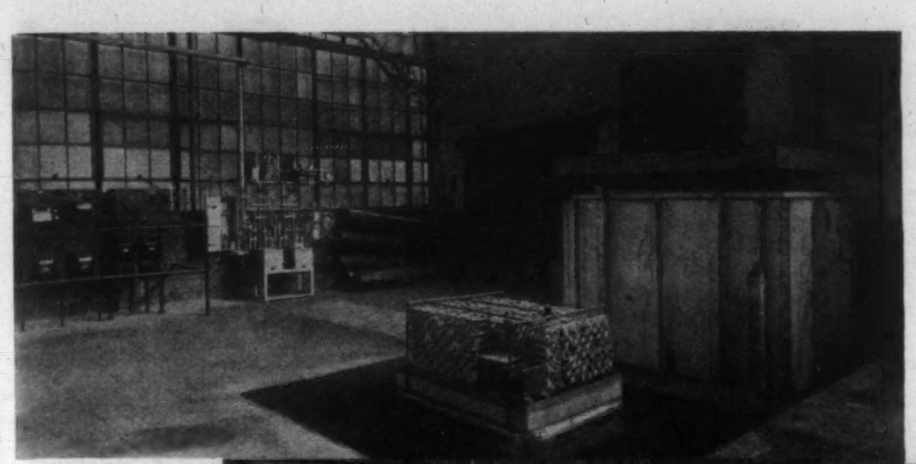

Fig. 4.

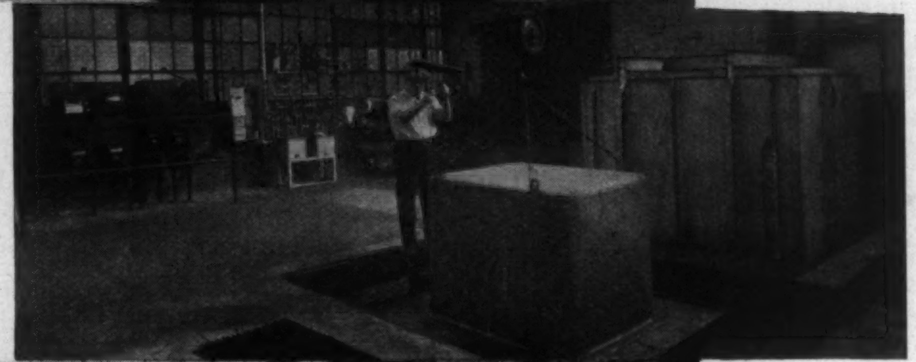

Fig. 5.

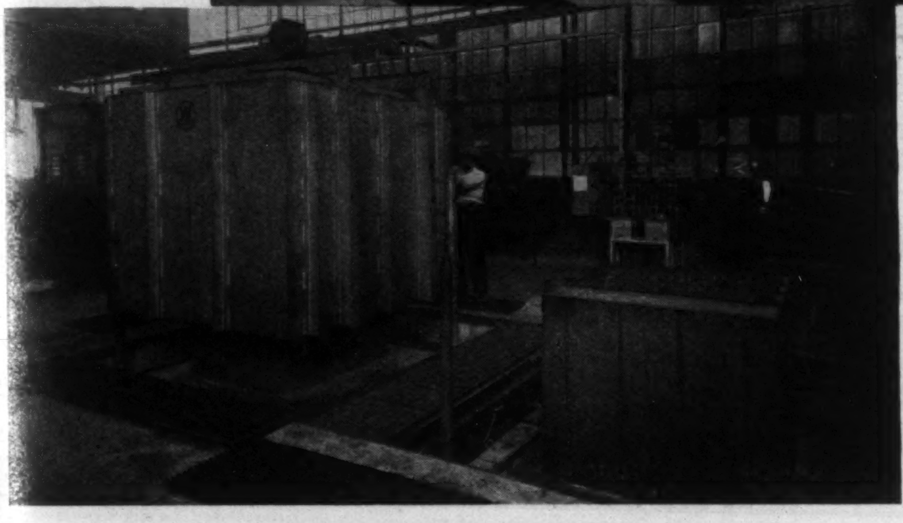

Fig. 6.

installations at both the branch Works at East Springfield and South Philadelphia.

To insure the best materials and immunity from wear in the mechanism of the Westinghouse Refrigerators, the nitriding plant was installed at the East Springfield Works almost entirely for that purpose. In Fig. 4 is shown a typical charge, of crankshafts and other refrigerator detail parts. In the background will be seen the supply and control apparatus similar to that already described in the East Pittsburgh plant.

In these newer plants there are however a number of markedly improved features. It will be noted in this illustration, Fig. 4, there is no container but that the work has been placed in racks and stacked on a base grid. Fig. 5 shows the hood lowered over the stack of work and Fig. 6 the "bell furnace" being lowered over the hood. It will be appreciated in this setup the work is more easily placed and removed and that the heavy box container and lid have been supplanted by a grid base and thin shell hood, of either nickel or Monel metal. Again for hermetically sealing the work the lead seal of the box container has been replaced by an oil seal at the base of the outer bell.

Like all new industrial processes this one is continually developing and finding wider scope of application. Nitrided sheets of 1/8" thickness upward are being employed for a number of commercial applications and numerous other parts now are being fabricated, pressed and deep drawn from this material.

The application of nitriding up to this time has been limited in some fields as it has been considered that where alteration to nitrided parts has been desirable, this could not be done without denitriding or softening the whole detail. It is now however possible to soften for machining any local spot or area without materially affecting the hardness of the surrounding case. This is of considerable importance in connection with the manufacture of die casting dies, etc. It is sometimes desirable to relocate core pins, vents, or make other adjustments which hitherto could not be done without softening the entire tool and subsequently renitriding.

Again high speed steels are receiving considerable attention from a nitriding aspect and already excellent results have been obtained with drills, form tools, etc., and as much as 10 times the life compared with the normal tool is being obtained. Advantage also is taken of the characteristic "growth" and worn tools on nitriding or renitriding are brought back to their original dimensions. This procedure alone has effected considerable economy.

Nitrided parts at elevated temperatures as high as 550°C. are little affected and retain their maximum hardness and general physical characteristics whereas a case-hardened case under similar conditions would be considerably softened.

Certain alloy cast irons have proved responsive to the nitriding process giving a Vickers Brinell hardness of around 300. This again offers further industrial application in the direction of gears, internal combustion engine cylinders, etc.

This new and important field of the metallurgical industry is yet in its infancy and not unlike the stainless steel industry will quickly develop as its many great advantages over the carburized case become more apparent to the engineer. Attention to the present has been largely concentrated on details machined from bar stock, small stampings and forgings. Apart from the still greater fields of application in that direction attention now is being given to the application of this process to castings of suitable composition in both steel and cast iron.

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Dr. Alcan Hirsch, chief chemical engineering advisor to the Russian Government is returning to Russia. He will act as special correspondent to METALS & ALLOYS and will keep our readers informed on metallurgical progress in Russia.

Samuel Frankel, formerly Director of Sales for the Niagara Falls Smelting & Refining Corp., has been appointed manager of the Special Alloys Department of H. Kramer & Co., Chicago, smelters and refiners of metals.

Earl Steward, formerly associated with the Albert Pack Corp., has been appointed Western District Manager of the Molybdenum Corp. of America, with headquarters in the First National Bank Bldg., Chicago, Ill.

Frank L. Gibbons, formerly with the Republic Steel Corp., has been appointed Director of Steel Sales, Timken Steel and Tube Company, Canton, Ohio.

Electric Steel

By HENRY D.

PRACTICALLY all of the output of acid electric furnaces engaged in steel production is utilized in the making of steel castings. Therefore any discussion of this particular means of producing steel must take into consideration problems peculiar to this industry. It is not intended herein to propound any new theories in respect to steel making practice, but rather it is desired to set forth some empirical observations made while engaged in the production of acid electric steel for castings.

The manufacture of the metal is a matter of vital concern to the steel founder, since he is not able, as is the producer of mill steel, to eliminate or ameliorate the damaging effects of certain defects, such as cavities, or inclusions, by mechanical working. One of his main concerns, in common with other manufacturers, is the problem of increasing output and decreasing percentage of rejections. He is properly appreciative of the necessity for the absence of defects, such as mentioned, in the interiors of casings to be subjected to severe service conditions, particularly those involving dynamic stress.

His problems are far from being as nearly ideal as the ingot producer's; exigencies of design make it a more complex one. Instead of having, as does the ingot producer, a symmetrically designed mold of correct taper, which aids in the progression of solidification along definite lines, as well as a dry, clean, metallic mold which, when in proper condition, can give off no gases and which offers great resistance to the erosive effect of the metal, he has to produce castings, differing in section and practically never symmetrically designed and properly tapered, in a mold composed of non-metallic material which gives off, as a rule, large volumes of gases and vapors and which is, even under the best of conditions, fairly easily eroded as compared to the ingot mold.

The ingot producer having a nearly ideal set of conditions may take care of liquid shrinkage by the imposition of a head at exactly the correct location and need not be bothered by cracks formed during solid contraction as no resistance to this is offered by the mold. The steel founder, however, confronted by complex designs, must concern himself with allowances for liquid and solid contraction under adverse conditions.

Therefore, with all the incidental barbarisms imposed on the steel by the average sand mold, a steel of originally high quality is necessary if the resulting casting is to show high service efficiency.

Inclusions in Steel

Inclusions, or impurities mechanically associated with the metal at ordinary temperatures are now receiving considerable attention.

There are 3 major classes of inclusions to be found in steel and they are:

1. The oxides (including silicates, aluminates, etc.)
2. The sulphides.
3. The oxide-sulphides.

In the production and handling of acid electric steel there are 6 possible sources of inclusions:

1. Non-metallic matter in or on the charge.
2. Non-metallic matter formed upon use of a recarburizer as a result of the oxidation of silicon, manganese and aluminum and by the introduction of sulphur.
3. Formation of non-metallic matter upon improperly working the heat with ore or by allowing severe reduction of silicon from the silicious slag to take place.
4. Formation of non-metallics when the final additions of silicon, manganese, aluminum, titanium, zirconium, etc., are made.
5. Formation of non-metallics due to faulty brick-work, spout or ladle.
6. Formation of non-metallic matter within the mold due to sand erosion.

The last source of inclusion formation does not lie totally without the steel-maker's domain, for molds in condition to produce satisfactory castings have served as prolific sources of inclusion formation when poured with cold or sluggish metal of proper cleanliness and analysis.

With no sulphur and oxygen present there could form neither sulphides nor oxides and hence there would be no inclusions. At present it is not economically feasible to use materials free from sulphur nor is it possible, with present steel making processes based upon oxidation, to dispense with oxygen. Consequently, sulphide and oxide inclusions will be present. To recapitulate:

Fundamental sources of inclusions are:

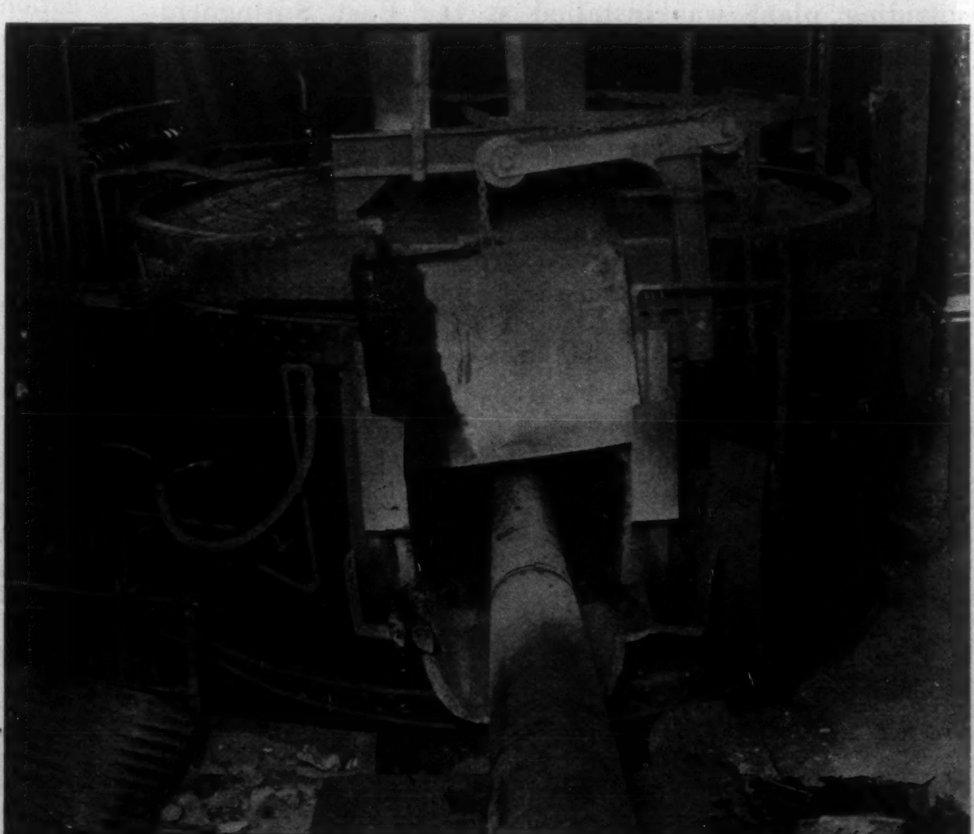
1. Oxygen } Introduced into the metal bath through the
2. Sulphur } charge and furnace atmosphere.

Subordinate sources are:

1. The slag in contact with the metal.
2. The refractories in and about the furnace.
3. The mold.

Inclusions tend to form definite points of weakness in the otherwise strong, ductile metal, particularly when present as elongated shapes or localized aggregates.

By applying the theory of elasticity to the problem it can easily be seen how inclusions form points of weakness in the metal. The edges of inclusions are regions of high stress concentration, thus explaining why cracks are often found to start from them when the metal is subjected to alternating stress. Once a crack is started,



Close-up of electric furnace being charged by the charging machine, showing the ram about to dump the contents of the charging box into the bath. (Timken Roller Bearing Company.)

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* Metallurgist, West Lawn, Pa.

For Castings

PHILLIPS *

stress concentration is increased, facilitating further cracking until failure results.

The presence of inclusions is especially harmful, therefore, in brittle and hard metals in which cracks are easily started. With the increased interest in, and production of, alloy steel castings, some of which are inherently brittle as cast, they assume a position of major importance. Small globular inclusions are least harmful, while elongated or continuous shapes, as well as localized aggregates, are most dangerous.

Kinzel and Crafts[1] concluded from their observations: that the dynamic strength of steels is quantitatively reduced by visible inclusions or factors directly proportional to them; that the magnitude of their effect on the dynamic strength increases with the hardness of the steel; that factors other than visible inclusions have an important effect on the dynamic strength, and are comprised in the term "body"; that this "body" is markedly influenced by minute inclusions whose size is less than 0.005 mm.

This term "body" has long been used by steel makers and seemed to imply a sum total of those things about which they knew little or nothing. The importance of these unknown factors have been demonstrated, however, in numerous instances.

There seems to be quite a divergence of opinion regarding the status of inclusions in the molten metal. Most metallurgists are inclined to the belief that inclusions are insoluble in the liquid metal. Ziegler,[2] however, believes that inclusions are not only soluble in the molten metal but also in solid γ-iron. McCance[3] and Giolitti[4] believe that inclusions exist as suspensions in the iron.

Wohrman[5] states that a considerable proportion of the inclusions found in iron and steel form prior to solidification and exist in the melt as suspensions, while an equally (if not more) important proportion is formed upon solidification, from materials dissolved in the molten metal. He also states that no inclusions of importance are formed after solidification and that inclusions are fixed in the solid metal and do not migrate.

Sims and Lilliequist[6] state that the normal inclusion-forming material is soluble in the molten steel and precipitates at solidification. Their observations lead them to attach considerable importance to the influence of iron oxide on the form of the precipitated inclusions.

Herty[7] states:

> "Experimental work has shown that FeO is soluble in liquid and solid steel. It is well known that sulphur is quite soluble in liquid steel and less soluble in solid steel. Although we have no direct experimental evidence on the solubility of MnO or SiO_2, we know that these two materials must be present in solution in liquid steel to a limited extent. The fact that the reactions
>
> $$Mn + FeO \leftrightarrow MnO + Fe \quad (1)$$
>
> and
>
> $$Si + 2\,FeO \leftrightarrow SiO_2 + 2\,Fe \quad (2)$$
>
> are both reversible, that in the basic open-hearth the amount of residual manganese may be calculated from a consideration of the equilibrium of reaction, and in the acid open-hearth that the residual silicon may vary according to concentration and temperature changes in the system, prove that these two oxides must be soluble to some extent in liquid steel and therefore to a more limited extent in solid steel. When steel solidifies, a great deal of the soluble material is thrown out of solution, but some portion remains in solid solution."

As long as methods of steel production hinge upon oxidation it will not be possible to totally prevent the formation of inclusions, but it is possible to limit their formation by wisely directing furnace and ladle operations, and perhaps substituting for such conventional deoxidizers as silicon, manganese, aluminum, etc., a deoxidizer whose oxidation products have a strong tendency to escape from the metal or remain in the form of tiny globules evenly distributed.

Abnormal Behavior of Some Acid Electric Steels

Acid electric steel made under certain conditions exhibits abnormal behavior at times. One striking example is "over-reduced" steel, being metal from which, according to Batty[8] an excess amount of oxide of iron, or suboxide of iron, has been removed, due to extravagant reducing conditions in the furnace. Another example is the great discrepancies existing at times between steels of like orthodox analysis, in respect of conventional physical properties.

"Over-reduced" steel has been described by Batty as lacking in fluidity, though satisfactory in chemical composition as ordinarily determined, and in temperature. This condition can be remedied by an addition of iron oxide to the bath, hence the steel is logically designated as "over-reduced."

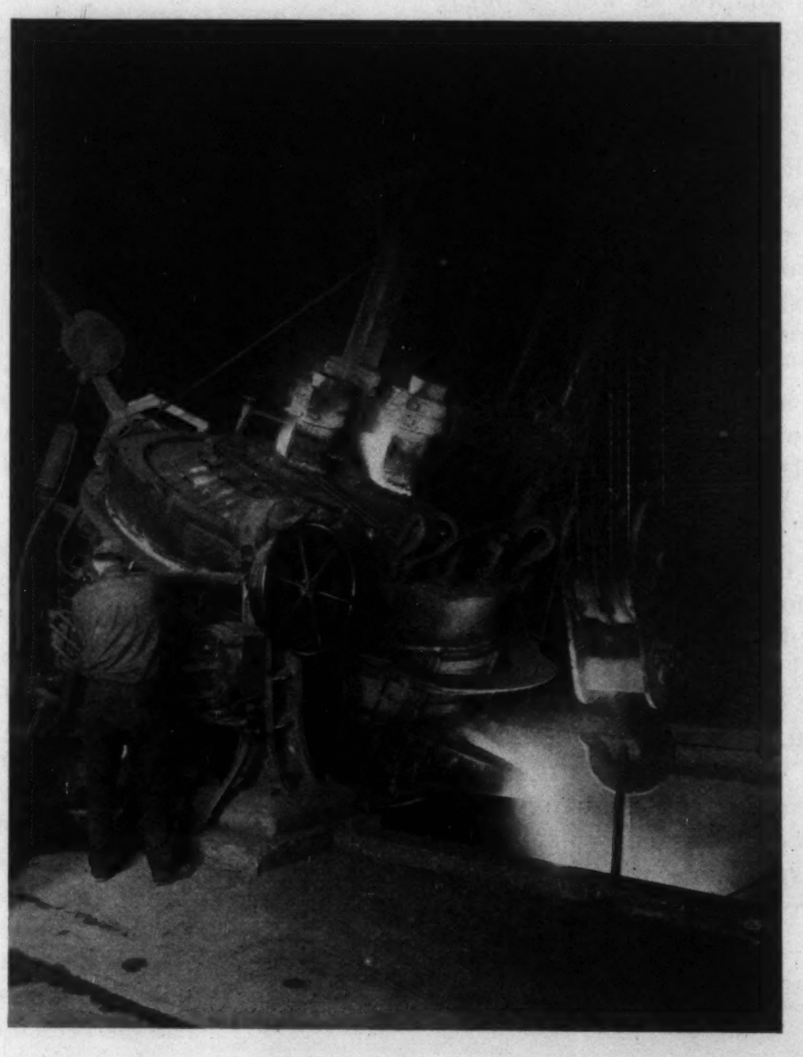
In the usual acid electric steel practice the melter's count test is used as a measure of fluidity. The best method of applying this test is to pour a sample of the steel into a standard open core-sand mold, noting the number of seconds required for the sample to film over. This is a much more accurate indication of fluid life than can be obtained by allowing a spoonful of metal to set in the spoon, for in the acid practice the slag may become more unmanageable than in basic practice, and as a result the spoon may become heavily coated with slag, thus holding a smaller amount of metal, or may become

Tilting "Teapot spout" ladle pouring direct into sand moulds. (Courtesy Pittsburgh Electric Furnace Corp.)

coated with a light slag covering and hold a larger amount of metal. Varying quantities of steel in the test spoon are apt to wrongly affect conclusions.

These core sand molds are of a capacity somewhat less than the spoon and are provided with a spill gate, so that samples of uniform size are obtained. A sample of metal is taken from the furnace in a spoon and poured into a core mold, any excess metal flowing out through the spillgate. Once the spoon is emptied time starts and the seconds are counted until a film completely forms over the surface of the metal in the core. In a highly fluid metal small specks of film appear and move about on the surface, some being re-absorbed and later reappearing. In metal of low fluid life, a film quickly forms over the surface of the sample and although, at times, a fairly satisfactory time test may be obtained otherwise, this quick forming film may be taken as an indication of a condition unsuitable for the production of light steel castings.

Many steel melters consider fluidity to be a direct function of temperature and conventional composition. The error of this belief is quite evident to one who has had any experience with over-reduced steel.



Tapping 6-Ton Electric Furnace. (Courtesy Bethlehem Steel Company.)

Batty concludes from his observations that fluidity (or fluid life) in a steel is related to the presence of dissolved oxide or suboxide of iron and that a steel containing little or none of this dissolved oxide evolves a great deal of its surface heat as light and hence loses its fluidity. He found that these steels are lighter in the liquid condition and have a higher freezing point than a steel otherwise similar in composition but containing normal amounts of dissolved oxide of iron.

This presumed presence of dissolved oxide of iron in mild steel for castings results in increased productivity from metal in the ladle due to the almost entire absence of freezing of metal in the nozzle (or handshanks), and the reduced size of the risers which may be employed. It results in fewer rejections due to the slightly lower solid contraction, the ability to produce thinner sections and the better surface of the castings resultant upon the decreased surface tension of the more fluid steel contacting with the mold during casting. Freer machining properties may also be noted. Of particular importance to the furnace operator is the decreased wear on refractories and reduced power consumption because the requisite fluidity is attained in a shorter period, and apparently at a lower temperature.

To quote Melmoth[9] on the subject:

> "Over a large number of electric furnace casts it is strongly suggested that abnormally high silicon produces a material much more difficult to cast from small shanks than when silicon content is normal. The principal point observed is the tendency to skin over the surface of the metal immediately after it is exposed to the air. It appears incorrect to state dogmatically that this is due to the direct action of the presence of silicon, as it may be equally due to the furnace condition which tended toward the production of the silicon content. These charges are frequently very dead and result from strongly reducing conditions. Even when hot they do not exhibit the usual degree of fluidity and have been termed "dazed" heats. The condition is remediable by a return to oxidizing conditions in the furnace, afterwards finishing as usual."

When tests of normal steel are made before and after the additions of the finals it will be found that the use of silicon and manganese has reduced materially the fluidity of the metal. Considering that there has been only a slight drop in temperature upon the additions of these finals, along with an appreciable increase in carbon and manganese content (both of which elements should increase the fluidity of the metal) it is found hard to correlate the two unless it is ascribed to some obvious constitutional difference other than the modification of carbon, manganese and silicon contents. It is significant, however, that in an over-reduced steel the addition of the finals does not seriously lower the fluidity and in some cases actually increases it. As manganese and silicon react with the iron oxide present it is to be assumed that the constitutional change consists of a lowering of the iron oxide content. It is not possible to prevent this depreciated fluidity in the steel, but it is possible to prevent the low fluidity condition brought about by severe reducing conditions in the furnace, when nascent silicon enters the metal. When this condition does arise, reversion to normal fluidity may be secured by an addition of oxygen to the charge.

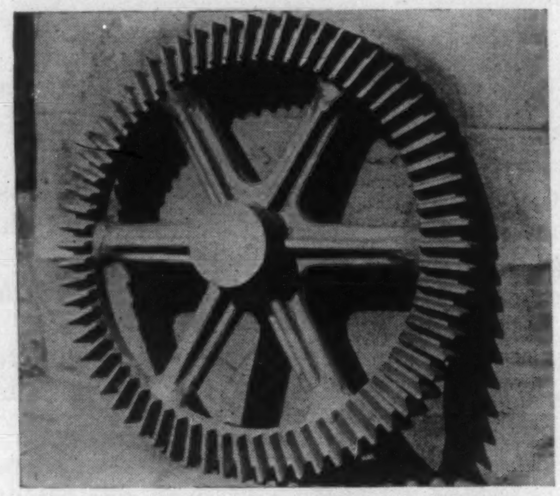
In the production of high chromium steels in the acid furnace this same condition evidences itself. It appears that under certain conditions such steel is especially prone to be over-reduced.

Foundrymen who have had experience with both electric and converter steels claim a greater "life" for the latter. Considering the procedure followed in making converter steel it can easily be seen, if the over-reduced theory is considered, why this should be so, and why they do not need such large risers for converter steel as for electric steel.

This greater fluidity is ascribed by some to be due to the completion of a reaction while the metal is in the ladle, between the manganese, silicon and oxygen contents. This may be true to a certain extent, but it does not explain all the phenomena attending an over-reduced condition.

Many steel makers have noted an abnormal behavior of some steels produced in acid and basic electric furnaces, when it was impossible to kill the metal. Fortunately such a condition is rare as an attempt to deoxidize the metal by any known medium meets with no success. Solidification and re-melting appear to remove the trouble. At times the furnace samples of such steels do not indicate their character; being sound but upon being cast into molds they have risen badly. At other times it is impossible to secure a quiet furnace test. The silicon content in either case is usually high, occasionally as high as 0.8%. Aluminum, manganese, zirconium and calcium have no effect. It appears to be due to some abnormal association of the silicon content due to improper furnace conditions.

An example of this condition follows:—Having secured satisfactory tests on a normal carbon steel, the heat was tapped. It behaved nicely in the ladle but upon being poured into the molds rose considerably, it being impossible to fill the molds completely. Samples taken from several castings analyzed 0.20% C, 0.40% Mn and 0.30% Si, whereas the last sample before tapping analyzed 0.26% C, 0.65% Mn and 0.48% Si. It is quite probable that these steels were loaded with free, active oxide of iron. Comparisons of the analysis indicate this.

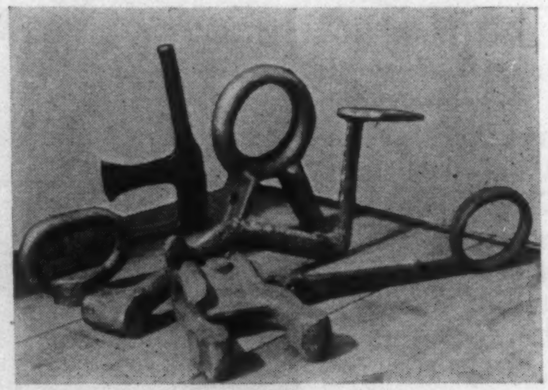


Steel to produce this casting must be of excellent fluidity or "runability" in order that contours and dimensions of teeth be accurate.*

When some acid electric steels are compared according to the formula given below there are great discrepancies existing at times between steels of like composition.

$$\text{Factor} = \frac{\dfrac{\text{Yield Point lbs./in.}^2 + \text{Max. Stress lbs./in.}^2 \times \text{Elong. \%}}{2}}{100 - \text{Reduction of Area \%}}$$

These same discrepancies exist between some acid electric and acid open-hearth steels. Melmoth[10] conducted an investigation of these phenomena (which seem illogical when one considers that the steels are of equivalent composition and produced under comparable slags, the only difference apparently being in the method of heat application) and concluded that the blame should be placed on the presence of ultra-microscopic silica inclusions in the electric steel unless carefully produced and that the electric process tends toward such a development especially because of the relative time in the furnace.



A group of castings, of no particular intricacy, demanding free-running metal without such excess of temperature as would tend to promote the injection of mold gases at internal angles.*

This is further evidence of the importance of those influences implied by the term "body."

EFFECT OF ELEMENTS PRESENT IN ACID STEEL

In plain carbon steel for castings variations in conventional composition may influence the behavior of the steel as cast into molds. It has been noticed that variations in carbon content, aside from affecting the purely physical properties of the metal at atmospheric temperature, may influence the ability of the metal to withstand the damaging influence of mold gases. It seems that very low carbon steels have a great capacity for mold gas absorption. This is probably intensified by the fact that their high solidification temperature necessitates a high casting temperature. It is also more difficult to produce absolute freedom from porosity in a low carbon steel. Such steels are lacking in fluidity as well.

Silicon in steel is apparently present in solid solution in the metal. It is primarily present to promote the soundness of the castings by causing gases to be retained in solid solution during solidification and to cause the deoxidation of the steel by forming silicon dioxide, SiO_2. Being held in solid solution, it would not be expected to have any effect upon crystalline form. This seems to be the case when steels of 0.3% silicon and one free from this element are examined, but it does not seem to hold when higher silicon contents are reached. It seems that high silicon, 0.4% or more, under certain conditions tends toward the formation of a coarse ferritic structure with consequently lowered strength and shock resistance properties. It may also render the metal unable to retain the normal amount of gases in solid solution or may tend, perhaps, toward an abnormal gas absorption of the metal in the mold. The metal not being able to hold this gas upon solidification, it is liberated with blowhole formation. High temperature with high silicon content offers the best opportunity for this temporary absorption. It should be emphasized that silicon variations should be avoided as far as possible in the production of light steel castings in green sand molds, the successful pouring of which requires a high temperature.

It is known that manganese promotes the soundness of steel castings through its efficient deoxidizing action in liquid steel. It seems to assist soundness very greatly in green sand castings due to its apparent ability to enable the iron to retain dissolved gases at solidification, its action differing in this respect from silicon which apparently increases such solution capacity while liquid, some gas being evolved at solidification. Manganese is also beneficial as a result of its action upon the sulphur present in iron, combining readily with this element to form manganese sulphide either alone or combined with iron sulphides. It appears from late investigations that it is mostly combined in varying proportions with FeS. This manganese tends to cause the sulphides to occur as rounded inclusions rather than as a network around the primary grains, and thus aids in producing a stronger, more ductile steel.

The two elements sulphur and phosphorus usually present in steel are known to exercise damaging influences though the extent to which these influences are felt when neither one is present to more than 0.05% has been, as a rule, exaggerated by most foundrymen.

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*Courtesy Steel Castings Development Bureau.

On the whole it may be stated that variations in any of the elements normally present in steel should be avoided and therefore it is desirable that a method of production should be used which renders us more nearly immune from these variations.

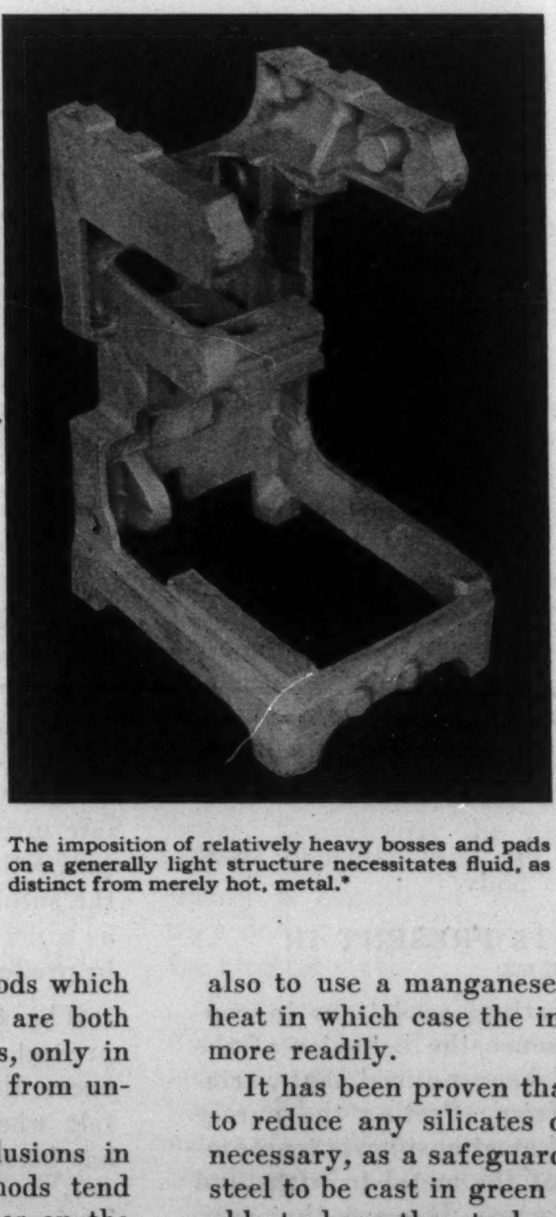
Steel Making

With increased knowledge of slag significance there has been marked improvement in steel making in any and all of the various processes. Whereas it was usual to hear of the "art" of steel making until comparatively recently, it can be stated today that the physical chemistry involved is gradually being understood and controlled. True, while some of the reactions are incompletely interpreted, some of the more important are well understood. The origin and nature of inclusions as well as the composition, and effect of gases in the steel are subjects which merit and demand further investigation.

Regarding the methods of steel production in the acid electric furnace to be discussed, some of the theories advanced may be wrong but their application is of commercial importance in producing steel free from abnormal behavior and otherwise unsatisfactory properties.

To discuss first the so-called "open-hearth" method of steel production it is found necessary to add a higher carbon material to the average charge so that the carbon content at melt down may be in the neighborhood of 0.35%. The carbon content is then lowered by the action of the slag, small additions of iron ore or limestone being added at times in order to liberate enough free iron oxide from the slag to bring the carbon down to the 0.18 or 0.20% desired. In this case temperature, slag and metal all converge to one point, at which time the finals are added and the heat tapped.

An alternative method consists of so composing the charge that it melts down at about 0.20 or 0.22% carbon. The carbon is then eliminated until only about 0.12 to 0.15% is present. This elimination requires an excess amount of FeO in the bath which then functions in the fluxing of any silicious non-metallics introduced with the charge by the formation of Fayalite (or ferrous ortho-silicate). Having this excess oxide of iron present in the bath and not wishing to kill it off with silicon, with the resultant formation of finely divided silica rich inclusions, wash metal is added to the bath, the carbon of this metal reducing the oxide of iron to form carbon monoxide which causes a boil and passes off. When the slag has reached the right consistency and a carbon estimation is made an addition of wash metal is again made to raise the carbon to the desired figure of 0.18 or 0.20%. This addition is allowed to react and the bath come to comparative rest when the finals are added and the heat tapped.



The imposition of relatively heavy bosses and pads on a generally light structure necessitates fluid, as distinct from merely hot, metal.*

There is quite a difference in opinion regarding the best method to be used in the production of the steel. While the two methods which have just been described seem to differ they are both fundamentally sound and really differ, it seems, only in minor points as they both produce steels free from undue inclusion content and abnormal behavior.

Considering the 6 possible sources of inclusions in acid electric steel it is found that both methods tend toward the elimination of inclusions added in or on the charge, with, of course, the sulphides excepted. In the first method a slow boil with a slag of good fluidity, necessarily rich in basic oxides, favors the coagulation, levitation and attachment to the slag of any silicious particles. In the second case a sustained boil to a low carbon content, which necessitates a high iron oxide content and consequently favors the formation of fusible silicates, insures the same elimination.

It is known that in order for inclusion material to coagulate it must be of low surface tension which necessitates a fairly low fusion point as compared to steel. In order to levitate with any speed it must coagulate into an appreciably good size and once levitated it must attach itself to the slag to prevent being again washed into the bath by convection currents ever present. With low surface tension as in these cases, it flattens out at the slag-metal interface and is then absorbed.

The second cause is also eliminated as the use of wash metal, which is high in carbon and low in manganese, silicon, sulphur and phosphorus is advocated as a recarburizer in one case, while in the other method no recarburizing additions of this sort are needed.

The third cause is also under control as any ore added to the charge acts as a flux or reacts with carbon, the carbon monoxide being formed leaving the field of action as a gas and aiding in bringing the fluxed inclusions to the slag.

It might be well to reserve the fourth cause until the fifth and sixth are briefly considered. The fifth cause is the result of mechanical failure and therefore not concerned in the physical chemistry of the process. The remedy is obvious. The sixth cause lies within the molder's domain if satisfactorily fluid steel, said fluidity not obtained solely as a function of temperature, is delivered to the mold. Such fluidity is possible in steels produced by these methods.

To revert to a consideration of the fourth cause. It is obvious that with the slag rendered fairly inactive and slag and metal nearly in equilibrium in respect of iron oxide there will be a minimum formation of inclusions upon the additions of the finals. If the manganese and silicon are added together, their speeds of reaction with iron oxide not being greatly dissimilar, any inclusions formed being rich in basic oxides would tend to coagulate, levitate and attach themselves to the slag. It is possible also to use a manganese-silicon alloy in deoxidizing the heat in which case the inclusions formed seem to levitate more readily.

It has been proven that aluminum added to steel tends to reduce any silicates or other oxides present. As it is necessary, as a safeguard at present, to add aluminum to steel to be cast in green sand molds it is obviously desirable to have the steel as free as possible from such inclusions upon adding the aluminum.

A procedure which is common in some European open-hearth plants and which may be suggested here is the use of MnO in the acid slag. Average acid slags made without extra MnO analyze at finish about as follows: FeO 23%, MnO 15%, CaO 4%, SiO_2 55%, and with MnO used FeO 11%, MnO 27%, CaO 4%, SiO_2 55%.

Another suggestion involves the use of lime in acid slags. It is known that it is possible to secure a steadier carbon elimination when lime is added to the slag than when no lime is added. An addition which will give a CaO content of about 10% is quite safe.

Both MnO and CaO tend to replace FeO in the slag and it is thus possible to lower the FeO content without lowering the slag's effectiveness in carbon elimination during the working of the heat. At the finish of the heat there is likewise a lesser liability of FeO diffusion after deoxidation.

There is no liability to over-reduction or abnormal behavior in these steels as no opportunity is allowed for a considerable reduction of silicon from the slag. On the whole it may be stated that the abnormal behavior associated with acid electric steels are eliminated with perhaps the exception of the variations in the effect of aluminum or alumina upon the steels.

Cable Sockets*

In the normal production of steels by these methods there is a greater working time of about 10 or 15 minutes in the making of an average steel foundry heat, namely 3 tons. It is obvious to one who has observed the benefits resulting from this however, that the extra trouble and cost in steel making is offset several times in the foundry.

Use of Aluminum

It is necessary at present to add aluminum to steel used for green sand casting in order to enable the metal to withstand the unbalancing effects of mold gases.

It is generally known that aluminum added to a steel containing FeO will react with this oxide to form finely divided alumina which tends to form localized aggregates, thus militating against the development of its best physical properties. Interest has been evidenced of late regarding the effect of an aluminum addition upon the sulphide arrangement in the steels so treated. It seems that aluminum additions, under certain conditions, cause the sulphides to be arranged in a streaky network structure as compared to the coarser, globular arrangement usually found in steel not so treated. This network structure causes a decided depreciation in the ductility of the steel and is therefore especially deleterious.

The fact that this condition may be found in steels not treated with aluminum would indicate that this element is not directly responsible.

Sims and Lilliequist[6] state,

> "Accompanying this change in the inclusions is a decrease in ductility, principally in the reduction of area, although the elongation is also affected. Occasionally the addition of aluminum does not produce this characteristic change in the inclusions, and when it does not, the ductility is likewise unaffected. Sometimes low ductility is found in steels not treated with aluminum, but in these steels is found a type of inclusion similar to those produced by the addition of aluminum."

They advance the theory that it is due to iron oxide content of the steel stating,

> "When aluminum is added to steel, it first reduces iron oxide. If it reduces enough of it, the sulphide inclusions will be changed from the globular to the eutectic type, and the globular iron-manganese silicates to the refractory aluminum silicates. If there is a large quantity of iron oxide present, or only a small addition of aluminum is made, enough iron oxide will remain to precipitate the sulphides and the steel will not suffer in ductility. If iron oxide is added to steel which has already been treated with sufficient aluminum to change the sulphides to the eutectic type, the sulphides again take the globular shape, * * * *. But in such a case, the iron-manganese silicates are almost absent.
>
> "As may be seen * * * *, a powerful deoxidizer, such as calcium silicide has the same tendency as aluminum to cause low ductility through its indirect effect on the sulphides. A high carbon content has a like tendency. In short, any condition that will lower the iron oxide content sufficiently will set the stage for low ductility. The obvious remedy is to maintain a proper amount of iron oxide by any method that will attain this end. Such a remedy has worked successfully in repeated tests."

It is known that oxygen decreases the solubility of sulphur in iron so the contentions of Sims and Lilliequist seem logical. This streaky sulphide arrangement has been noted in steels made under improper furnace conditions and not treated with aluminum, the removal of the necessary iron oxide being accomplished by atomic silicon reduced from the slag.

True iron oxide equilibrium between slag and metal is seldom reached even in average open-hearth practice because carbon elimination is proceeding at all times preventing the metal from taking up iron oxide to its equilibrium value. Considering the relatively greater speed of operation in the electric furnace and the possibility of nascent silicon entering the field of action it may be seen that still less opportunity is allowed for the metal to attain an iron oxide content equilibric with its carbon content. If the slag, during the latter stages of the heat is low in basic oxide content reduction of silicon from the slag is bound to result. This introduction of nascent silicon to the metal decreases the FeO content, already low due to the highly acid slag. Therefore, if a bath of metal with an equilibric FeO content is desired the slag during the working of the heat should be kept high in bases such as MnO and CaO.

If a slow rate of carbon elimination is caused to exist and the reduction of silicon from the slag is prevented there is a greater chance of the metal attaining its FeO equilibrium. As silicon, manganese and aluminum effect this iron oxide content, lowering it, it would seem advisable to limit their presence.

As the manganese, due to its carbide forming property, and its effect upon the sulphides, has a beneficial effect it is advisable to allow its presence to exist in the proper average proportions. It does seem advisable, however, after considering the aforementioned effects of high silicon content, to limit its presence. Well made steel will produce sound castings with a silicon content as low as 0.20%. In the methods discussed it is possible to make good steel and produce sound castings with this silicon content. It is advisable, however, to aim for a silicon content about 0.25% as a precautionary measure. Well

(Continued on Page 232)

Percent Elongation as a Method of Measuring

By REID L.

Introduction

INVESTIGATION of the effect of shape of test specimen on the results of the tensile test date back at least as far as 1880 when Barba reported his classical experiments[1] and proved what has since come to be called "Barba's Law of Similarity." An explanation of this "Law of Similarity," together with some references to the work of other investigators, is given in several texts on strength of materials[2] and there are also some recent papers.[3] There are certain phases of the problem which did not seem to be fully answered by previously published work and these will now be discussed.

Nichols, Taylerson and Whetzel,[4] in 1927, published a number of conversion factors for comparing the percent total elongation on different shaped specimens of thin sheet metal. Such factors are very useful for comparing the results of different laboratories in which specimens of different dimensions are used. It was an attempt at such a comparison which led to the present investigation.

Sheets of low carbon steel of deep drawing quality were tested in tension using 2 different test specimens; one was ½ inch wide over a parallel section 2¼ inches long and the other was the same width over a parallel section 4¼ inches long. The gage lengths were 2 and 4 inches, respectively, and on both specimens the shoulders were 11/16 inch wide and the radius of the fillets approximately 1 inch. The percent elongation on the short specimen was found to be 1.12 to 1.15 times that on the longer one instead of the ratio of 1.23 reported by Nichols, Taylerson and Whetzel.[4]

On the long specimens the elongation was measured in the entire 4 inches and also in the 2 inch section containing the fracture and it was found that the ratio of these 2 checked that found by these investigators. There seemed to be 2 possible causes for this behavior, either one or both of which might be the explanation; difference in speed of testing, and effect of the fillets or shoulders on the elongation near the end of the parallel section.

Outline of Tests

A very complete study of the effect of variables in size and shape of specimen on the properties determined in the tensile test was reported in 1926 by Templin[5] but in these tests there was no variation in length of parallel section for a given gage length. The effect of testing speed is frequently dismissed by writers on this subject by stating that "it does not appear that such variations in the rate of loading as are liable to occur in practical tests of iron or steel have much influence on the extension or the strength." On the other hand, Bach and Baumann[6] show that speed may have an appreciable effect. In view of the different statements made regarding the effect of testing speed on the percent elongation it was considered advisable to include a study of this variable in the present investigation.

It is generally accepted that the percentage elongation is a function of both gage length and cross section. Long specimens emphasize the effect of gage length, while short ones give prominence to the local contraction which is dependent on the cross section. Different kinds of drawn forms present different problems, but there are many cases in which it is important that material for deep drawing withstand considerable elongation without local reduction. This is an argument in favor of a long rather than a short gage length. The fact that some investigators[7] favor an 8-inch gage length for measuring the ductility of drawing sheets indicates the importance attached to the "general" elongation, for the longer the gage length the more this factor predominates. It is possible to determine the "general" or "uniform" elongation on a sample of any gage length; see below.

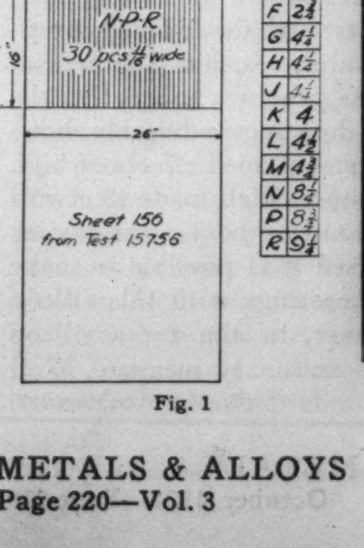
With all these points in mind, the investigation was divided into 4 parts which will be enumerated here and then separately described.

1. Study of the effect of testing speed on the relationship between the elongation observed on samples with 2 and 4 inch gage lengths.
2. Study of the effect of proximity of the shoulders to the measured gage length upon the observed percentage elongation.
3. Study of various methods of measuring uniform elongation and reduction of area and effect of gage length on these quantities.
4. Determination of elongation correction factors for specimens of different thickness, but of equal length and width.

For these tests a sheet of 19 gage, low carbon steel of the following analysis was used:

Carbon	0.06%	Sulphur	0.024
Manganese	0.30	Silicon	Trace
Phosphorus	0.010	Copper	0.047

This sheet was normalized, pickled, and



Set No	Parallel Section
A	2¼
B	2¼
C	2¼
D	2
E	2½
F	2¾
G	4¼
H	4¼
J	4¼
K	4
L	4½
M	4¾
N	8¼
P	8¾
R	9¼

Fig. 1

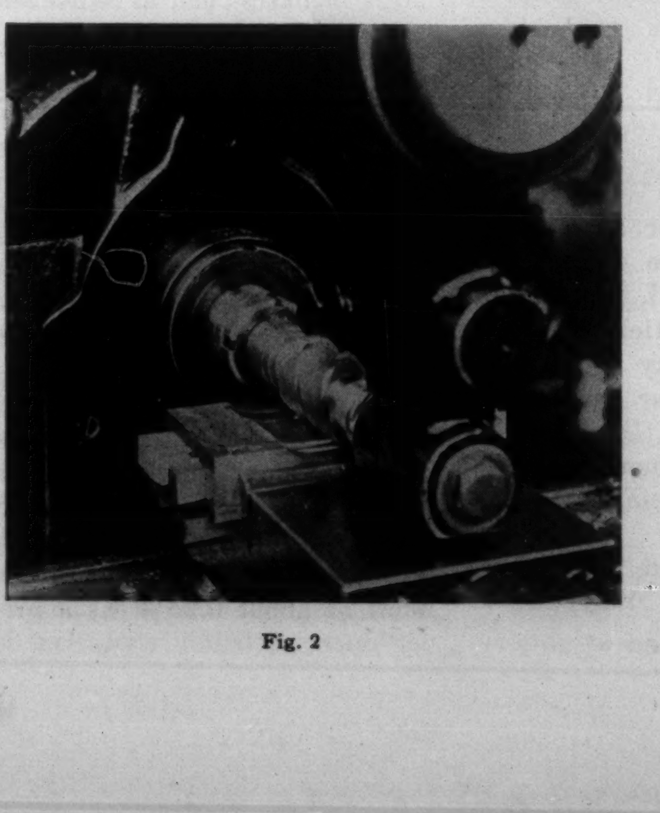


Fig. 2

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in the Tensile Test
the Ductility of Thin Sheets

KENYON†

box annealed and was from the middle of a cross rolled slab from the middle of an ingot. Other tests on material selected in this manner show it to be very uniform. The specimens were cut out as shown in Fig. 1. There were 10 samples in each group and all values given in this paper are each the average of a group of 10 tests. All specimens were prepared by milling sheared strips. The 2- and 2¼-inch parallel section samples were held in a special jig designed for machining these test pieces (Fig. 2). All other specimens of a given size were held together with tool-maker's clamps and supported on "parallels" in the vise, using "hold-down" clamps. Some of the samples had to be draw filed in order to maintain parallelism within 0.0005″ per 2 inch of parallel section. Fig. 3 shows the dimensions of the different specimens. All tests were made on the 2000 lb. range of a 20,000 lb. Amsler universal testing machine, using Templin grips for holding the specimens. A load-deformation curve was drawn for each specimen.

1. Effect of Testing Speed

The specimens were of the same shape as regularly used for the 2 inch and 4 inch gage lengths, i. e., 2¼ and 4¼ inch parallel sections, both being ½ inch wide. Thirty specimens of each kind were prepared and 10 were tested under each of the following conditions:

a. One set of 2″ gage length specimens tested at a head-speed of approximately 0.175″ per minute.*
b. One set of 2″ gage length specimens tested at a head-speed of approximately 0.35″ per minute.
c. One set of 2″ gage length specimens tested at a head-speed of approximately 0.70″ per minute.
d. One set of 4″ gage length specimens tested at a head-speed of approximately 0.35″ per minute.
e. One set of 4″ gage length specimens tested at a head-speed of approximately 0.70″ per minute.
f. One set of 4″ gage length specimens tested at a head-speed of approximately 1.40″ per minute.

The proper setting for the speed control valve was determined in each case by preliminary tests. The same valve setting was used throughout the entire duration of a given test so that the head-speed was approximately the same above and below the yield point. Speed measurements were made by timing each test with a stop watch. The total elongation over the gage length was divided by the time in minutes, thus expressing this value in inches per minute. In order to reduce all speed measurements to a common basis, regardless of length of specimens, this was converted into inches per minute per unit of gage length. This latter value is really *rate of strain*, which is much more significant than speed of head of the testing machine. In order to compute the proper cross head speed for pulling samples of different gage length at the same rate of strain, it is necessary to take into consideration the length of the *parallel section* (not the gage length). As an illustration, a 4 inch gage length specimen with a 4¼ inch parallel section would not require quite twice the cross head speed as a 2 inch gage length sample that had a 2¼ inch parallel section, in order that both could be tested at the same rate of strain.**

The results of the tests outlined above are given in Table I, and are plotted in Fig. 4, from which it can be seen that speed has a definite effect on the quantities

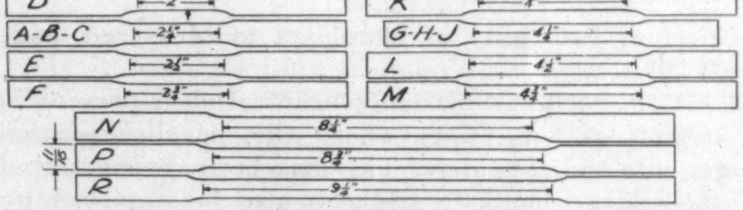

Fig. 3

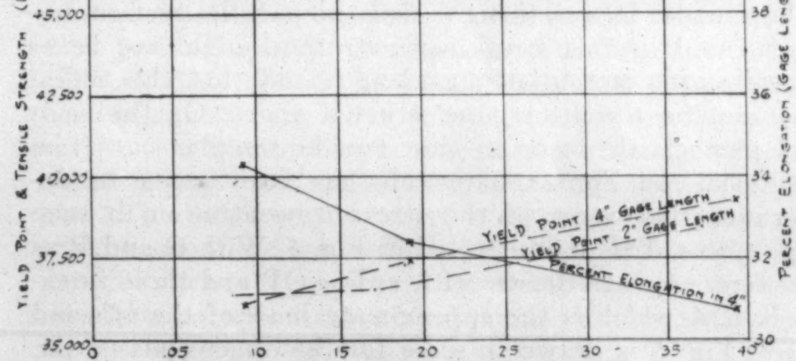

Fig. 4

Table I.—Effect of Testing Speed on Physical Properties

Samples	Rate of Strain in./in./min.	Yield Point lbs./in.²	Tensile Strength lbs./in.²	% Elong. (total)
2-¼″	.083	36,360	48,940	41.85
parallel	.170	36,580	49,235	40.6
	.371	38,430	49,850	38.6
4-¼″	.091	36,025	48,980	34.3
parallel	.192	37,440	49,235	32.4
	.392	39,280	49,710	30.75

Table II.—Ratio of Percent Total Elongation in 2″ vs 4″ When Samples are Pulled at Different Speeds

	4″ Gage Length		
2″ Gage Length	Slow*	Medium	Fast
Slow	1.22	1.29	1.36
Medium	1.18	1.255	1.32
Fast	1.12	1.19	1.255

* Rates of strain are those given in Table No. I.

Table III.—Effect of Length of Parallel Section on Percent Elongation

Parallel Section	Rate of Strain in./in./min.	% Elong.	
		%	Gage Length
2″	.165	39.05	2″
2-¼″	.170	40.6	
2-½″	.185	40.8	
2-¾″	.198	39.75	
4″	.168	33.15	4″
4-¼″	.192	32.4	
4-½″	.184	33.15	
4-¾″	.194	33.0	
8-¼″	.169	30.19	8″
8-¾″	.185	30.08	
9-¼″	.204	29.74	

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*Actual testing speeds are given in Table I, expressed as rate of strain.

**For studies of the effect of speed of tensile testing at stresses up to the yield, the rate of straining cannot be accurately regulated—owing to the small amount of strain, slippage in the grips, elastic deformation in the testing machine, etc.; the Section on Speed of Testing of Committee E1, A.S.T.M., therefore wisely chose the rate of stressing as a measure of speed of testing since their interest lay in the effect of speed of testing up to the yield point, but for greater elongations, it would appear better to measure the rate of straining, for reasons given in the present paper.

measured in the tensile test. Faster speeds decrease the percent elongation and raise the yield point. All of the yield points are based on the "lower yield" as read from the autographic diagrams which were drawn for all tests. It is also to be observed that increasing the speed raises the yield point slightly. When the ratio of percent elongation in 2 inches to that in 4 inches is computed, the results shown in Table II are obtained. These figures show that when the samples are pulled at the same rate of strain, the ratio of elongations is about the same, regardless of what that rate is. In the paper by Nichols, Taylerson and Whetzel, the ratio of 1.23 is given for converting percent elongation in 4 inches to that in 2 inches (on ½ inch wide specimens). There is no statement in this paper as to the testing speeds used, but Taylerson has informed the author that the various specimens were pulled at approximately the same rate of strain. Their value of 1.23 checks reasonably well with the ratios given in Table II, of 1.22 – 1.255 and 1.255. It is important to bear in mind, however, that the ratio changes considerably if the rate of strain is different on the 2 specimens being compared and in ordinary practice it is quite likely that the tests would be made at the same speed of head rather than the same rate of strain, especially if a testing machine is used that is provided with only a limited number of speed changes.

There seemed to be 2 possible causes for the discrepancies observed between ratios actually found and those reported by other investigators in the case cited in the introduction. These results indicate that the difference in rate of strain was sufficient to account for the discrepancies observed.

2. Effect of Proximity of Shoulders to Measured Gage Length

A certain amount of stress concentration is present in the region of the fillets where the parallel portion merges into the shoulders. This has been demonstrated by photoelastic methods and can also be observed by watching the formation of Lüder lines or "stretcher-strains" during the tensile testing of a piece of dead soft annealed, but fine grained high finish mild steel sheet. If the cross section is uniform over the gage length these lines will form first at the fillets showing that the stress intensity is higher there than in the parallel portion of the sample. It was thought that the parallel section, uninfluenced by this stress concentration, might not be as long as the measured gage length and that this might account for the discrepancies noted above. On the basis of photoelastic tests on thin tensile samples cut from celluloid, an approximate rule has been given to determine the extent of the stress concentration in such specimens. This is illustrated in Fig. 5. With O and P as centers, arcs are drawn with radius OP and these intersect at A which is the approximate limit of the affected area. Fig. 5 is drawn to scale for the dimensions of the standard 2 inch gage length test specimen and it can be seen that this indicates a slight encroachment on the parallel gage length section.

Table IV.—Effect of Rate of Strain on Uniform Elongation (Specimen Having a 4¼" Parallel Section).

Rate of Strain in./in./min.	Percent Uniform Elongation
.091	23.5
.192	22.8
.392	21.6

Table V.—Effect of Length of Parallel Section and Gage Length on Uniform Elongation.

Parallel Section	Gage Length	Rate of Strain in./in./min.	Percent Uniform Elongation
4	4	.168	21.19
4-¼	4	.192	22.80
4-½	4	.184	23.16
4-¾	4	.194	24.18
8-¼	8	.169	24.75
8-¾	8	.185	25.11
9-¼	8	.204	24.06

In order to carry out this part of the investigation, ½ inch wide specimens were prepared with parallel sections of the following lengths:

2" — 2-¼" — 2-½" — 2-¾"
4" — 4-¼" — 4-½" — 4-¾"
8-¼" — 8-¾" — 9-¼" —

Ten specimens of each of the above dimensions were tested at speeds calculated to give, as nearly as possible, the same rate of strain. There was greater variation in these rates than desirable, due to an error in the method of computation of the head speed. The values as given are correct rates of strain. The variation in these rates for the different kinds of test pieces is probably not enough to affect these results seriously.

The average results are given in Table III together with the rates of strain. Except for the specimens with the 2-inch parallel section, the length of the parallel section does not seem to have much effect on the percent elongation for samples of the same gage length. In other words, excepting the 2-inch specimens, the proximity of the shoulders seems to have no appreciable effect on the percent elongation. While this seems to contradict the rule given above, it should be stated that this is only approximate and that it errs on the side of safety. Evidently the 2¼ inch parallel section is long enough to prevent the shoulders from affecting the elongation of the measured gage length of the standard specimen.

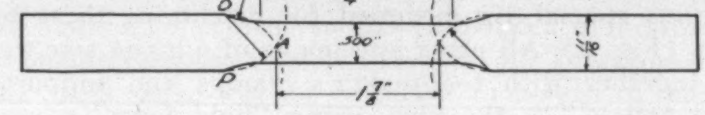

Fig. 5

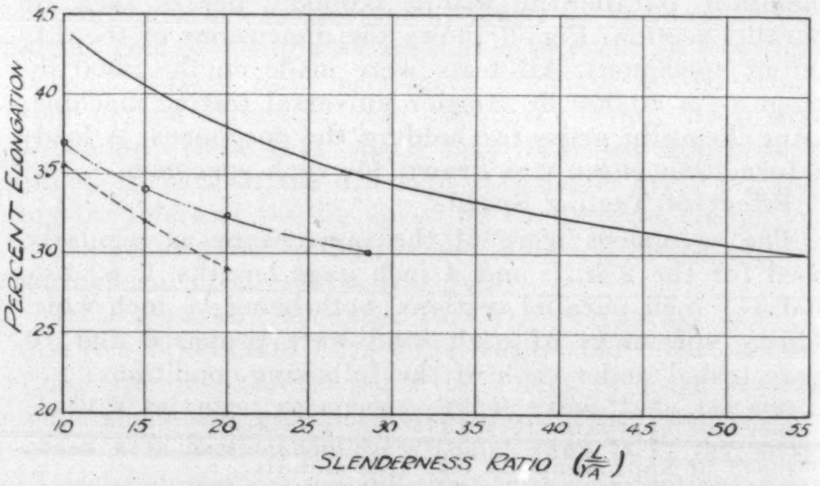

Fig. 6

3. Uniform Elongation and Reduction of Area

The total elongation of a tensile sample is the sum of 2 different effects: The general elongation over the entire gage length and the local elongation at the fracture. The first is directly proportional to the gage length, while the latter is a function of the cross sectional area of the sample. The relative effect of these 2 factors can be varied considerably by changing the gage length relative to the cross sectional area, or to express the thought more precisely, the ratio of the gage length to the square root of the cross sectional area $L/\sqrt{A}$, the so-called slenderness ratio.

It is the total elongation after fracture that is usually reported and this is influenced by the "slenderness ratio" as shown in Fig. 6. The solid curve gives the results of this investigation while the dotted curve presents the results reported by Nichols, Taylerson and Whetzel.[4]

The dot-dash curve is from Templin's results[5] on soft aluminum. It will be noted that all of these curves show the same tendency, i.e., to become asymptotic to some minimum value as the "slenderness ratio" increases. This is characteristic of the total elongation value. As has been said, the total elongation after fracture is the quantity generally reported and the reason for this is that it is more readily determined, but it is not the most significant measure of ductility as far as drawing quality is concerned. There are many cases in which deep drawing sheets must have the ability to elongate as much as possible without "necking-in." This is only another way of saying that the uniform elongation must be as high as possible, because up to the "necking-in" point, the material stretches uniformly over the entire gage length. Kuntze[8] and others have demonstrated this point and have also called attention to the significance of this value. Apparently the uniform elongation is truly characteristic of the material while the total elongation is not, due to the disturbing effect of the shape of the specimen.

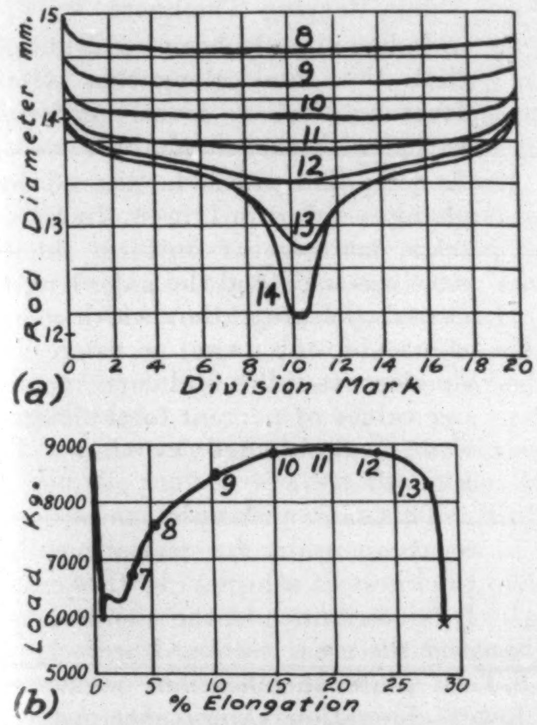

Fig. 7

Granted the significance of the uniform elongation it still remains to develop a means of determining it in practice, a problem which is attended with some difficulties. As Timeshenko[9] states that it is difficult to determine with accuracy the amount when necking begins and to establish separately the magnitude of the uniform stretching and the magnitude of the elongation due to necking. Morley[10] agrees with this and points out that it would form a suitable criterion of ductility were it not so troublesome to measure. He states that it cannot be measured satisfactorily after fracture, as the contraction at fracture influences the ultimate extension some distance from the fracture, the metal "flowing-in" towards the necked-in portion.

There is another more important reason than that given by Morley for not measuring the uniform elongation on the fractured specimen, which is that the entire length of the test piece continues to elongate even after necking begins. Fig. 7 reproduced from the work of Kuntze and Sachs,[8] shows this very plainly. The curve marked "11" in Fig. 7a shows the diameter measurements of a round test bar at 20 equi-distant points along the parallel section at the instant that the maximum load was reached as shown by point "11" on the stress-strain curve, reproduced in Fig. 7b, which was made for this same test bar. Although the load decreases from this point on, it is seen that the diameter of the test bar decreases considerably all along its length before fracture occurs. The various contour curves in Fig. 7a are numbered to correspond to certain points on the stress-strain curve shown in Fig. 7b. The measurements of the broken specimen, therefore, cannot be used to determine the uniform elongation.

It is sometimes considered that, on sufficiently long specimens, there are portions some distance from either shoulder or fracture which are uninfluenced by stress concentrations at these points and that therefore these portions have stretched an amount corresponding to the uniform elongation. If this were true, the measurement of the cross section at such points or the increase in length of a short increment laid off before testing, would supply the data for calculating the uniform elongation. The work of Kuntze and Sachs shows quite definitely, however, that even the portions considerably removed from either fracture or shoulder have not elongated an amount equal to the uniform elongation, but have been stretched considerably beyond this amount. It is quite apparent, therefore, that the uniform elongation cannot be determined from any measurements taken from the fractured test specimen.

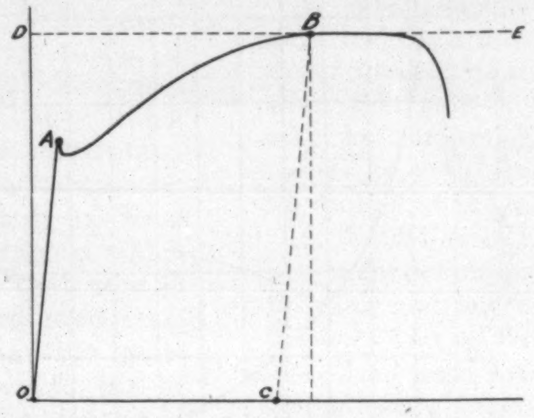

Fig. 8

If it were possible to determine accurately, during the course of the test, when the load first reached its maximum value, the corresponding elongation of the measured gage length could be taken as the uniform elongation. But it is quite difficult, if not entirely impossible, to do this and as the only remaining method is to make use of certain relationships between true stress and the cross sectional area of the test piece and determine the uniform elongation from the stress-strain curve as suggested by Vietórisz.[11] As the tensile test piece undergoes more and more plastic deformation, a point is reached where the decrease in cross section of the specimen just off-sets the cold-working or "strengthening" effect of the plastic deformation and the applied load reaches a maximum. Experiments already described have shown that the specimen elongates uniformly over the entire gage length up to this point at which necking begins. The point on the stress-strain curve to be taken is that at which the load *ceases to increase* because there is generally an interval in which the load holds constant due to the counteracting effects of strengthening and of decreasing cross section.

The method of obtaining the uniform elongation from the stress-strain curve is shown in Fig. 8. Ordinarily, the elastic portion of the curve is so nearly vertical that BC may be taken perpendicular to OC, but in any case OA and BC must be parallel to each other. The principal source of error in determining the uniform elongation, OC, is in the accurate location of point B, which is the point where the curve first becomes tangent to the horizontal line DE, through the maximum of the curve. An accurate stress-strain curve with the elongation magnified 10 times should give the uniform elongation by this method of measurement within ± 1%.

At the time of making the tests included in this investigation, the only recording 2-inch extensometer available was found to produce curves that were considerably too short compared with the actual measured elongation on the broken test pieces. This led to the development of an improved recording extensometer which will be described in a later paper.* In view of these facts, the uniform elongation could not be determined on any of the 2-inch gage length specimens in this series of tests. Such measurements were made, however, on the curves for all of the 4-inch gage length specimens tested at various rates of strain as reported in Part 1, and the 4 and 8-inch gage length samples of varying length of parallel section included in Part 2. The effect of testing speed is shown by the values given in Table IV. These show that the uniform elongation decreases with increasing rate of strain. A comparison of these results with those given in Table I indicates that the uniform elongation suffers a slightly smaller percentage drop for a given increase in rate of strain than the total elongation. In this case, an increase in rate of strain from 0.091 to 0.392 inches per inch per minute causes a 10.3% drop in the total elongation and only 7.3% for the uniform elongation.

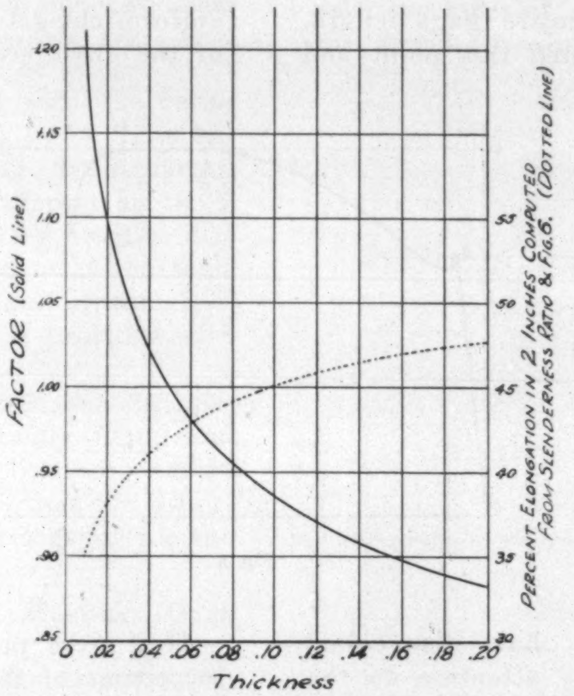

Fig. 9

The values given in Table V are for the 4 and 8-inch specimens of varying length of parallel section which were all tested at approximately the same rate of strain. These data bring out the 2 interesting points: first, that the length of the parallel section has some influence on the uniform elongation on the 4-inch gage length specimens, and second, that the uniform elongation is practically the same on both the 4 and 8-inch gage length specimens. In regard to the first point, it will be recalled that there was no significant difference in the total elongation on these same 4-inch gage length samples, although it is reasonable to expect that the longer parallel sections might give slightly higher values due to complete absence of the restricting influence of the fillets. The uniform elongation more nearly exhibits the expected relationship to length of parallel section, there being a spread of from 21.19 to 24.18 from the shortest to the longest specimens tested. The second point of interest is that the uniform elongation is practically the same on both the 4 and 8-inch gage length specimens. This confirms the statements previously made that this quantity is a characteristic of the material and not of the shape of the test specimen, providing that the gage length is free from the restricting effect of the shoulders at the ends of the specimens.

Reduction of Area

It is generally thought that reduction of area cannot be determined on sheet tensile samples due to difficulty in measuring the thickness at the fracture. However, the cross section of a few samples at the fracture was actually measured, using a binocular microscope with a micrometer eyepiece for determining the thickness. The results seemed to be quite dependable, although the method used for holding the samples for measurement was rather crude. With proper arrangements, this measurement could be made a part of the regular test. Although only a few thickness measurements were made, width measurements at the fracture were made on nearly all of the specimens tested. The averages of these measurements are given in Table VI, together with the ratio of width at fracture to original width, (W_3/W_1).

These results show that there is relatively little variation in the ratio of W_3/W_1 for the one type of steel tested, though the value might vary considerably from material to material; the usefulness of this value, however, is uncertain. The reason for introducing the above data is to show that the local reduction seems to be independent of the gage length, which is in accord with theory.

4. Correction Factors for Percent Elongation on Samples of the Same Width and Gage Length but Varying Thickness

It has already been brought out that the total elongation after fracture for a given material varies with the slenderness ratio, $L/\sqrt{A}$. This ratio is altered by changes either in L or A. In these tests, material of uniform thickness was used and the gage length varied, keeping the width constant. In this way, a range of slenderness ratios and corresponding values of percent total elongation was obtained as shown in Fig. 6.

This same principle can be used to compensate for variations in thickness of samples. In this case, L is constant and the slenderness ratio changed by changing the cross sectional area. In a rough way, all A.S.T.M. plate specifications make an allowance for the lower elongation values obtained on thinner plates of the same quality. For example, in several A.S.T.M. specifications (A7-29; A9-29; A113-29, etc.), a decrease of 1.25% is to be made from the nominal elongation requirement (22% in 8 inches) for each 1/32 inch decrease in thickness below 5/16 inches. For 5/16 inch specimens 1-½ inch wide, having a gage length of 8 inches, the slenderness ratio is 11.67. If the thickness is reduced to ⅛ inch, the ratio becomes 18.42, yet for this relatively moderate change in ratio, the elongation requirement is reduced from 22 to 14.5% in 8 inches. This is more than theory calls for.

If an agreement can be made upon some certain thickness to which all other thicknesses can be referred, the slenderness ratio can be used for correcting all elongations on different thicknesses of the same material to that for the "standard" thickness. For example, for

Table VI.—Reduction in Width of Samples at Fracture.

Set No.	Parallel Section	Gage Length	Rate of Strain in./in./min.	Average Width of Samples		
				W_1=orig.	W_3=frac.	W_3/W_1=ratio
A	2-¼″	2″	.170	.5007	.375	.749
D	2″	2″	.165	.5020	.385	.767
E	2-½″	2″	.185	.5003	.3825	.765
F	2-¾″	2″	.198	.5011	.3860	.770
H	4-¼″	4″	.192	.4988	.3765	.755
K	4″	4″	.168	.4996	.3805	.762
L	4-½″	4″	.184	.4839	.3665	.757
M	4-¾″	4″	.194	.5010	.3850	.768
N	8-¼″	8″	.169	.4988	.3845	.771
P	8-¾″	8″	.185	.4967	.3775	.760
R	9-¼″	8″	.204	.4995	.3785	.758

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*R. L. Kenyon & R. S. Burns. Autographic Stress-Strain Curves of Deep Drawing Sheets. Paper to be presented at the Buffalo Convention of the American Society for Steel Treating, Oct. 1932.

Table VII.—Effect of Variation in Thickness Upon Slenderness Ratio of (½″ x 2″) Tensile Samples and the Resultant Computed Effect on Percent Total Elongation.

Thickness t	Slenderness Ratio s	Percent Elong. in 2″ (From Fig. 9)	Factor Based on .050″ as 1.000
.010	28.285	34.75	1.212
.020	20.000	38.00	1.108
.030	16.331	39.75	1.059
.035	15.117	40.50	1.040
.040	14.142	41.10	1.024
.045	13.333	41.60	1.012
.050	12.650	42.10	1.000
.055	12.060	42.50	.991
.060	11.547	43.00	.979
.070	10.689	43.50	.968
.080	10.000	44.00	.957
.090	9.430	44.50	.946
.100	8.944	45.00	.936
.120	8.165	45.85	.918
.140	7.559	46.25	.910
.160	7.071	46.85	.899
.180	6.667	47.25	.891
.200	6 325	47.75	.882

sheets, 0.050″ might be taken as a "standard" thickness and all elongation values based on it for reference. It could then be said that material of 0.040″ thickness should only elongate 97.6% as much as the 0.050″ thick material (based on the kind and quality of sheet material used in this investigation). Since the slenderness ratio can vary due to change in either L or A, it has been possible to work out a series of conversion factors which should be applicable to material similar to that used in these tests, i. e., normalized, box annealed mild steel sheets. Similar relationships should hold for other materials such as brass, aluminum, and other metals, but of course the actual values for the correction factors would be different.

The conversion factors for the mild steel sheets are shown in Fig. 9, plotted against thickness. In applying these, the percent elongation found is multiplied by the proper factor for that particular thickness of material. The result will be the theoretical "equivalent" elongation which 0.050″ thick material of the same intrinsic ductility should exhibit. For illustration, the following example may be taken: Material 0.035″ thick, showed by test, an average elongation of 30.5% in 2 inches. The factor for converting to "equivalent" elongation for 0.050″ thickness is 1.040 so,

$$1.040 \times 30.5 = 31.7$$

On the above assumptions, therefore, this material is of the same intrinsic ductility as 0.050″ material which shows 31.7% in 2 inches on a sample of the same shape.

The method of determining the correction factors will now be described. In Part 3, Fig. 6 was presented to show the variation in total elongation with slenderness ratio. A range of slenderness ratios was obtained experimentally by varying the gage length of the specimens. Assuming a constant gage length of 2 inches, this range of slenderness ratios was converted into a range of corresponding thickness values. The percent total elongation can now be plotted against thickness instead of slenderness ratio (it is assumed from this point on that the gage length is always 2 inches and the specimen ½ inch wide). The dotted curve in Fig. 9 shows such a relationship for the material tested. This curve gives an elongation value of 42.10% for a thickness of 0.050″ which has been taken as a "standard" thickness. The values corresponding to other thicknesses can be read from the curve and these have been tabulated in Table VII (third column). The correction factors given in the fourth column of this table are merely the quotients obtained by dividing the elongation values for various thicknesses into 42.10. These factors can then be plotted against thickness as shown by the solid curve in Fig. 9. It should be said again that these values are applicable only to sheets that are similar to those used in this investigation, but the method should be useful in determining similar sets of factors for other materials.

It might be well to point out again that uniform elongation is not affected by the length of the specimen or the cross section. It is the local contraction only (which in turn affects total elongation) that is influenced by variation in cross section. Inasmuch as uniform elongation seems to be a better measure of drawing quality than total elongation, it would seem to be advisable to make the tensile test in such a way that this quantity can be determined. The uniform elongation, unaffected by variations in thickness and gage length is a more reliable quality factor than total elongation, for which correction must be made when different thicknesses or different materials are to be compared.

Conclusions

1. When specimens of mild steel sheet of different gage lengths are tested at the same rate of strain, the ratio of total elongation is about the same, independent of what that rate is. Percent total elongation in 4 inches may be converted into percent total elongation in 2 inches by multiplying by 1.24, providing both samples are pulled at the same rate of strain.

2. Variation in rate of strain from 0.08 to 0.38 inches per inch per minute raises the yield point about 2500 lbs./in.^{2} and lowers the percent total elongation, both in 2 inches and in 4 inches, about 3%.

3. Increased rate of strain lowers the percent uniform elongation.

4. Tests with specimens of different lengths of parallel section indicate that the 2-¼ inch parallel section is sufficiently long to give percent total elongation in 2 inches unaffected by the shoulders. It is also shown that the sample with a 4-¼ inch parallel section gives the percent total elongation in 4 inches unaffected by the shoulders. These conclusions refer to samples with 11/16 inch wide shoulders and ½ inch wide parallel sections and fillets of one inch radius.

5. Uniform elongation can be measured from accurate stress-strain curves. If the recording extensometer is accurate, the gage length is immaterial as far as the determination of this quantity is concerned.

6. The reduction of width of the samples at fracture seems to be independent of the gage length or rate of strain. This indicates that reduction of area at the fracture is also independent of these variables.

7. The percent total elongation is a function of the gage length and cross section, which may be expressed as a "slenderness ratio," $L/\sqrt{A}$. This relationship may be used to compare percentage total elongation on samples of the same cross section and varying gage length or on samples of the same length but varying cross section. Factors applicable to mild steel sheets have been determined for comparing the percent total elongation for different thicknesses of sheet, but the use of these necessitates the adoption of a "standard" thickness to which all others shall be compared.

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(Continued on page 232)

GAMMA-ALPHA TRANSFORMATION IN WROUGHT IRON

By R. H. HEYER*

MANY INTERESTING observations have been made on the 900°C. crystalline transformation of iron. Electrodeposited iron has been shown to recrystallize to an ordinary equi-axed grain structure at 910°-915°C.[1] The absence of recrystallization at A_2, 768°C., has been shown by microscopical and X-ray analysis. All evidence indicates that the region 900°-910°C. is the true temperature range for the γ-α transformation, the 768°C. thermal reaction being associated only with an inner-atomic magnetic change.

As early as 1895 the room temperature product of the $\gamma \rightarrow \alpha$ change in iron was proved to be somewhat more complex than the simple grain structure of most pure metals. Existence of more than one set of crystalline grains was pointed out at that time by Andrews[2] by means of microscopical examination of polished and etched wrought iron. The small grains appearing within the normal ferrite grains he considered to be due to a secondary crystallization.

The existence of two or more types of ferrite grain structures has been confirmed many times since and has been definitely connected with the polymorphic transformations of the iron in the later reports.[1, 3, 4, 5, 6, 7, 8, 9]

The commonest type of secondary structure has frequently been called α-veining. Alpha-veining was observed in restricted areas in Aston process wrought iron pipe, the material used in the present work. Typical cases of α-veining are shown in Figs. 14 and 15 taken from a specimen furnace cooled from 1000°C. The large, prominently bounded ferrite grains appear to be superimposed on a finer grained structure. H. B. Pulsifer[9] has appropriately given the name "phantom structure" to the sub-grain network. This puts the subject in the same class as "ghost lines" in steels except that the latter are somewhat more fully understood.

Many hypotheses have been advanced to explain the duplex structure found in some samples of ferrite. O. W. Storey[1] in his extensive work on electrolytic iron suggested an explanation for this structure which is widely accepted. He states that upon cooling from the γ region many α nuclei form in each austenite grain as the transformation temperature is approached. These nuclei have similar orientation within any austenitic grain, and after complete transformation the large α grains are subdivided into many smaller units arising from the similarly oriented nuclei. The inner grain boundaries are considered to be composed of more or less imperfect and very small α crystals of the same orientations as the surrounding metal. Such a material would tend to etch rapidly, but not as rapidly as the more thoroughly disorganized material comprising the boundaries between the large α units of different orientation. Storey observed that when 2 large adjacent grains had dissimilar orientation no continuity of the inner grains existed through the boundaries of the large grains. When 2 large adjacent grains had nearly the same orientation the inner grains were seen to cross the large grain boundaries.

The effect of mechanical work and annealing on α-veining has been investigated by Ammermann and Kornfeld[10] who conclude that veining may be eliminated by cold working, followed by heating below A_3 sufficiently long to recrystallize. Electrolytic iron so treated does not again become veined upon heating up to A_3. However, 15 minutes at 910°C. produced veining in this material, linking α-veining definitely with the γ-α transformation.

It is reasonable to believe that foreign bodies acting as centers of α formation have a definite, perhaps controlling effect on the transformation mechanism. Inclusion effects have been studied by Rawdon and Berglund[11] and by Bannister and Jones.[12] The ferrite of wrought iron is, of course, high in slag inclusions of microscopic size and may well be relatively high in sub-microscopic inclusions and soluble impurities, all of which may influence crystallization.

Rawdon and Berglund[11] by means of ordinary picric acid etching show 3 distinct types of veining in pure iron, 2 of which they associate directly with inclusions. The first of these networks they believe is caused by the localization of very small inclusions at the growth surfaces of δ iron grains as they form from the molten state at 1530°C. This δ network was found in cast material only and assumes the form under the microscope of stringers of inclusions in a relatively coarse, polygonal arrangement. This pattern was entirely independent of and superimposed on the true α grain boundary pattern. Aluminum-rich ferrite showed the δ network to a large extent, very likely because of increased inclusion content.

The same investigators found a γ network likewise appearing as a superimposed structure unrelated to the α grain boundaries. This network was not found in cast material but is similar to the δ network in that it is composed, at least partly, of inclusions. The γ network was so named because it appeared only after mechanical work was applied in the γ region or after long annealing in the same region. Quoting from page 713[11]: "Formation of this network implies some solubility of inclusions in the ferrite [γ ferrite] matrix and a decrease in solubility as the iron cools, or to some movement of the inclusions within the structure of the heated iron. It is very probable, therefore, that the inclusions which contribute to the formation of the γ network are more limited in number than those of δ network, the formation of which is assumed to be largely mechanical in origin as the crystals increase in size as they form in the molten metal."

Rawdon and Berglund[11] found the more familiar α-veining in both cast and worked metal. They definitely proved that the veining was not a secondary crystallization within the larger grains by applying an ammonium persulphate etch which colored the large units uniformly but differently depending on their orientation. From this they conclude that the veining grains are relatively undisturbed portions of much larger grains. In general, no close relationship was found between α-veining and the inclusions present.

Rawdon and Berglund[11] and Tritton[13] have further shown by means of slip and Neumann band observations that no orientation change exists within the large α grains.

Finally, the extensive work of Rawdon and Berglund brought out the fact that a transformationless ferrite, as produced by a 5% Si content, was not subject to veining either in the cast condition or after treatment.

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*Purdue University.

It is interesting to note that in 1910 Kroll[14] also found 3 networks in iron. His procedure in producing this structure, his Fig. 7, was to polish and etch, then heat slowly to 950°C. in a hydrogen atmosphere. The volume changes at critical points supposedly create 2 networks in addition to the original etched structure. The widest grooved network this author associates with γ iron because the greatest volume changes take place at its formation, and because of some evidence of twinning in these grains. The second network produced by heating appeared: "in relief as white streaks, which, in the author's opinion, forms the polyhedral boundary of the β iron."

The structures, Figs. 10 and 11, obtained by Kroll upon etching phosphoriferous iron (0.378% P, 0.03%C) for one hour with weak picric acid solution are more nearly similar to those obtained in the present investigation than any which came to the writer's attention in reviewing previous work (See Fig. 21).

The work of Bannister and Jones[12] requires further mention. They also found 3 structural types in iron. Their Type I structure consists of a small grained, heavily cored network within and independent of the α grains of mild steel and wrought iron. They associate this structure with phosphorus and develop it by deep etching with alcoholic nitric acid or cupric chloride reagents. The wrought iron used contained 0.19% P. Local segregation was suggested by increased amounts of coring in the vicinity of slag patches, in some of the larger α grains, and in the direction of rolling of longitudinal sections. Annealing in hydrogen for 5 hours at 820°C. eliminated the coring. The Type II structure of Bannister and Jones is the ordinary α-veining network. The Type III structure, produced by long etching in 0.5% nitric acid, has the appearance of etching figures both at grain boundaries and within. They consider Type II, or α-veining, to be caused by sub-microscopical inclusions which give the appearance of continuous lines—the veining boundaries. As the magnitude of the inclusions present increases, Type III appears in which the boundaries are made up of individual etching figures, presumably caused by the inclusions.

EXPERIMENTAL RESULTS

In the present investigation a short section of ¾" Aston process wrought iron pipe was quenched in cold water from 900°C. giving the structure shown in Fig. 4. (Alcoholic picric acid applied for a comparatively short time is the etch used in all of the micrographs except Figs. 12 and 17.) Around the groups of larger slag particles and at grain boundaries a new grain formation has appeared. This new structure was proved to be slightly elevated from the groundmass. A new device was used to establish this fact definitely.* By means of a double filter, green on one half and red on the other, and with central arc illumination, red light is cast obliquely onto the specimen from one side and green light from the other. This gives an effect as in Fig. 1. Elevation AD, corresponding to the special grains in Fig. 4, appears green at the left boundary and red at the right. Depressions such as the slag hole *BC* or the grain boundary *E* have the reverse order, red at the left and green at the right. The original determination of color order for depressions vs. elevations was made by observing the color effect on the raised carbide particles in annealed high speed steel. It is to be further noted in Fig. 4 that the grain boundaries of the large original ferrite grains penetrate the smaller, slag including grains. Another specimen cooled in air from 900°C., Fig. 5, has approximately the same amount of this special structure but the distinctness of outline has diminished. In this figure the focus was somewhat diffent from Fig. 4, the special areas no longer being in relief, in which position they are rather more indistinct. The difficulty in showing the special structure properly in relief (Fig. 4) and still showing the original grain boundaries distinctly (Fig. 5) is apparent. Most of the structures of the series were focused as in Fig. 4.

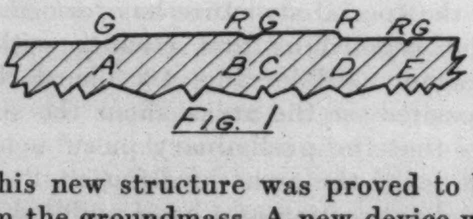

FIG. 1

Upon heating for one hour at 900°C. and air cooling, structures were obtained having approximately the same proportions of special grains as in Figs. 4 and 5.

Upon quenching a specimen from 910°C. after 15 minutes heating, the structure of Fig. 6 was obtained. Here the characteristics of Fig. 4 are more clearly portrayed and the special structure is greater in extent. From this start a series of specimens was heat treated at 10°C. intervals from 880° to 970°C. and at 1000°C. A general survey of Figs. 2 to 9 indicates that a gradual formation and retention of the special structure occurred between 890° and about 950°C. The coincidence of this temperature interval and the γ-α transformation temperature suggests a relationship between these structures and the crystalline transformation. Fig. 2, obtained by quenching from 880°C., showed none of the special structure but consisted of ordinary α grains whose boundaries did not etch as readily as in "as received" or annealed specimens. The deeper etch required, colored some of the grains. It is to be noted that the few special areas of Fig. 3 surround slag patches and should represent the first ferrite to transform, while the small patches in Fig. 9 are isolated from slag groups and should represent the last ferrite to transform to austenite.

It was noted that a small band at the inner circumference of the pipe sections was relatively free from slag and seemed to lag in the transformation from the old structure to the new. Fig. 9, taken after quenching from 950°C., is slightly higher than average in the amount of insular areas of the old structure remaining since many regions on this specimen were entirely free from them. At 960°C. none of the old structure remained. At 970°C. no further change was observed. Also at 1000°C. no further structural change was observed at 100X except perhaps an increase in grain size.

In addition to this series some supplementary tests and additional photomicrographs were made. A specimen was heated to 970°C. for 15 minutes, dropped back to 910°C. for 15 minutes, and then quenched. The amount of special structure obtained was about the same as in Fig. 6, quenched from 910°C., but the sharpness of the boundaries could not be brought out quite as well as in the latter structure.

Another specimen was heated to 920°C. and slow cooled. A normal α grain structure resulted instead of the duplex structure of Fig. 7, obtained after quenching from 920°C.

Fig. 10 was taken at 300X from the same specimen as Fig. 8. Here the elevation of the special structure is

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*Developed by Dr. Welch and Mr. Villela, Union Carbide and Carbon Company.

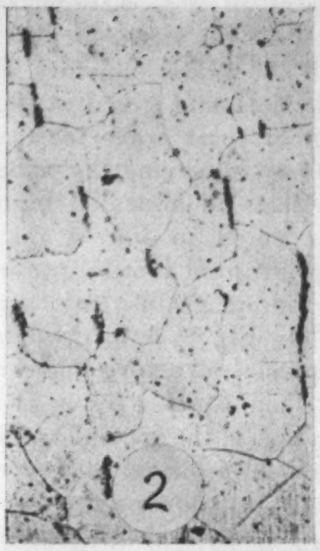


Fig. 2. 880°C. Water Quench. 100X.

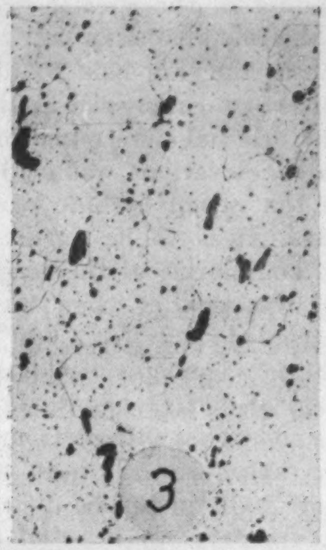


Fig. 3. 890°C. Water Quench. 100X.

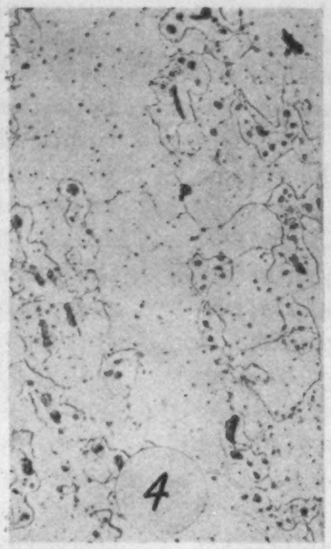


Fig. 4. 900°C. Water Quench. 100X.

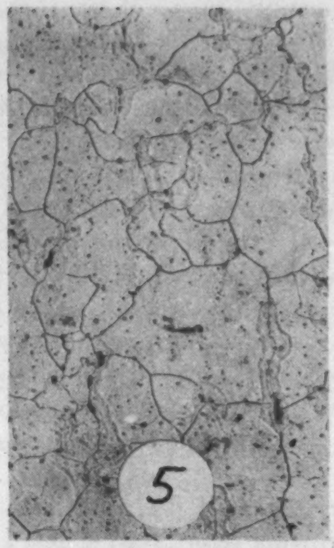


Fig. 5. 900°C. 5 Min. Air Cooled. 100X.

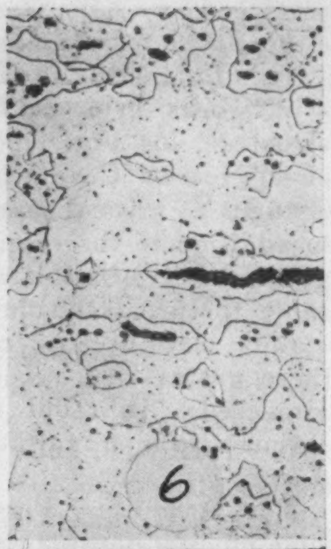


Fig. 6. 910°C. Water Quench. 100X.

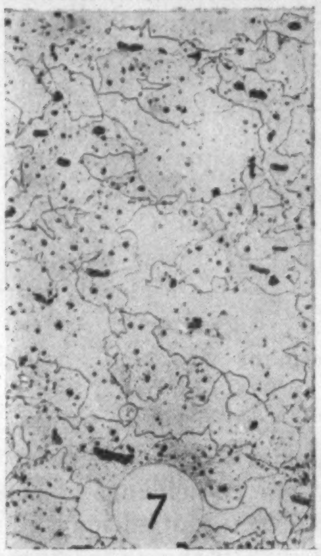


Fig. 7. 920°C. Water Quench. 100X.

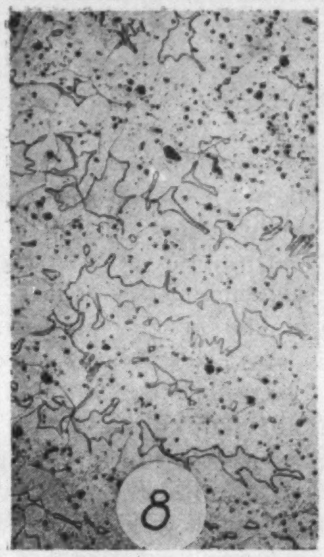


Fig. 8. 930°C. Water Quench. 100X.

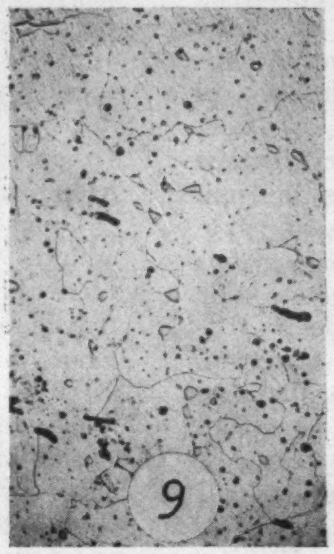


Fig. 9. 950°C. Water Quench. 100X.

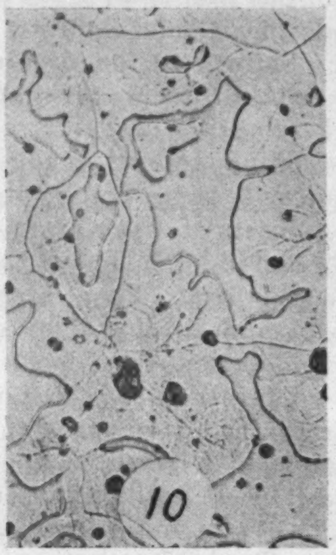


Fig. 10. 930°C. Water Quench. 300X.

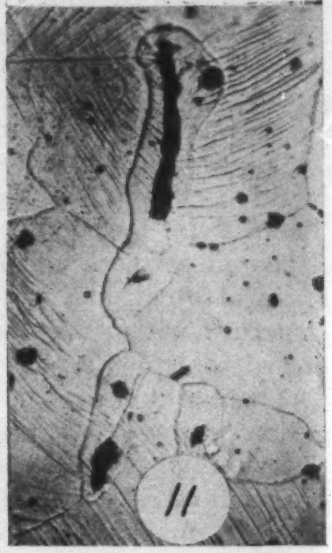


Fig. 11. 900°C. Water Quench. Stressed. 300X.

quite evident. The growth of this same structure about the larger groups of slag is also apparent.

Fig. 11 represents a specimen quenched from 900°C. and then squeezed in a vise to produce strain. Slip bands are seen to penetrate the 3 small areas of special structure without change in direction. This was found to be the general case. However, it was also observed that slip occurred much more frequently in the areas of original structure than in the special structure. This was true even in a specimen quenched from 930°C. and strained, and in which the special areas predominated the structure.

Fig. 12 shows the action of a cupric-chloride reagent (2 g. $CuCl_2$ – 1 cc. HCl – 50 cc. alcohol) on a specimen quenched from 910°C. The copper etch was preceded by the usual picric acid etch to define the boundaries between the 2 types of structures. In the microscope the double etched specimen appeared to be pure copper colored in the areas of special structure, and faintly copper colored in the vicinity of the grain boundaries of the unchanged parts. The central, uncolored portions of the latter grains formed bands in many areas. According to the work of Stead,[15] Rawdon,[16] and others, copper deposits first upon those areas having the lowest P content, therefore the special slag—including ferrite appears to be low in P.

Fig. 13 at 500X is a completely transformed area on the specimen quenched from 950°C. (See Fig. 9.) Deeper etching has revealed a double network as reported by Rawdon and Berglund[11] in their work on pure iron. The specimen quenched from 1000°C. also had a double network which was only very faintly visible at 100X. The former γ boundaries appeared to be rich in tiny inclusions when observed at higher magnifications.

A specimen slow cooled from 1000°C. had many areas of α-veining, see Figs. 14 and 15. The intensity of the veining network ranged from very faint, as in Fig. 15, to a condition where the veining network was equally as heavy as the larger grain structure. This condition was approached in Fig. 14. From numerous observations of areas as shown in Fig. 15 and of published micrographs, it was concluded that the α-veining net work was quite independent of the grain boundaries of the large α grains. With perfect polishing and uniformly deep etching the network boundaries should cross the boundaries of the larger grains more frequently.

In order to determine how general the foregoing transformation structures were, 4 other irons were examined: 1/4″ Aston process wrought iron plate was quenched from 910°C., giving the structure in Fig. 16 taken from a longitudinal section. Comparing it with Fig. 6 the grain size is seen to be smaller and a somewhat larger amount of the special structure has formed. Fig. 17 was obtained by repolishing and etching with the cupric chloride reagent used for Fig. 12. The dark streaks are copper deposited on the areas about the slag stringers. This shows that the preliminary picric acid etch is not required to define the areas of different P content.

A longitudinal section of a hand puddled wrought iron bar given a similar heat treatment showed the characteristic duplex structure. This wrought iron had been used about 40 years as a piston rod for a Corliss steam engine. It is designated here as 1890 wrought iron. An old wrought iron chain was also found to have the same type of structure after a similar heat treatment as shown in Fig. 18, and differs from the Aston process structures

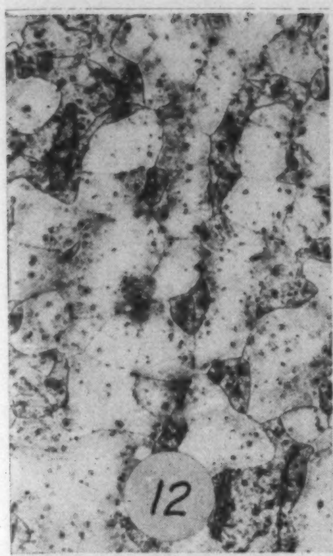


Fig. 12. 910°C. Water Quench. Etch: 1. Picric Acid. 2. $CuCl_2$, HCl. 100X.

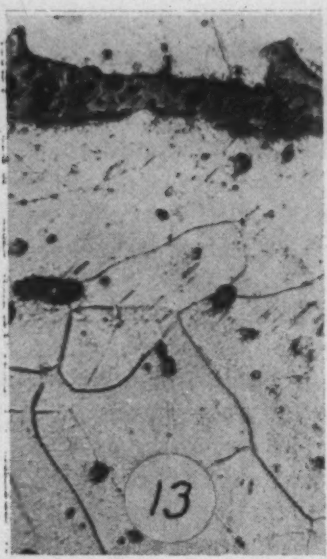


Fig. 13. 950°C. Water Quench. 500X.

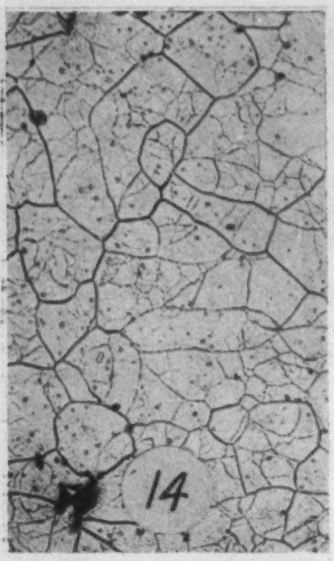


Fig. 14. 1000°C. Furnace Cooled. 100X.

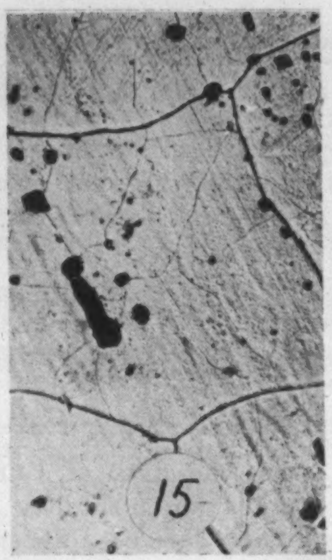


Fig. 15. 1000°C. Furnace Cooled. 300X.

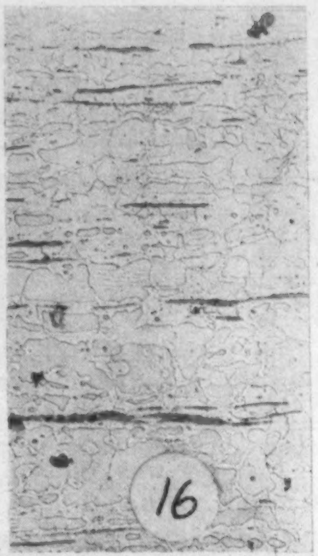


Fig. 16. Aston Wrought Iron Plate. Longitudinal. 910°C. Water Quench. 100X.

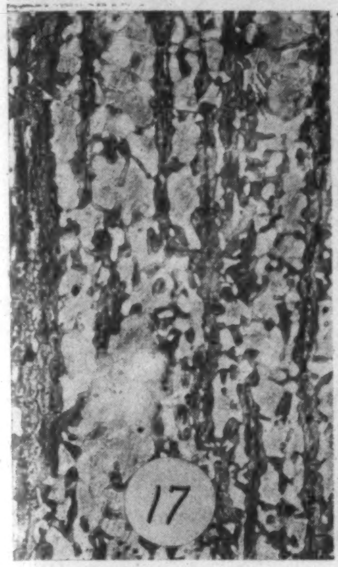


Fig. 17. ¼" Aston Plate. 910°C. Water Quench. Etch: 2g. $CuCl_2$, 1 cc. HCl, 50 cc. Alcohol. 100X.

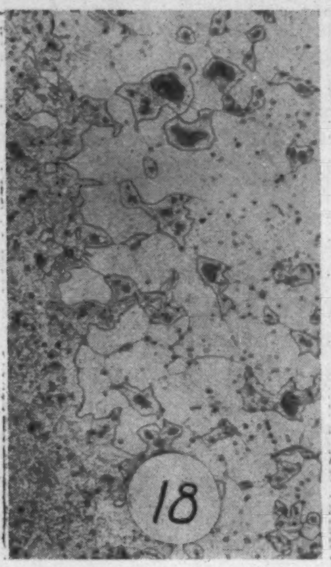


Fig. 18. 910°C. Water Quench. Hand Puddled Wrought Iron Chain. 100X.

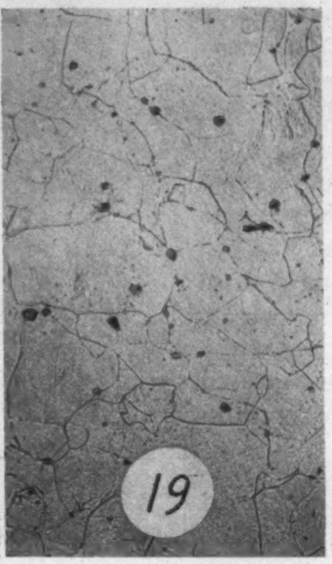


Fig. 19. Armco Iron. 900°C. Water Quench. 100X.

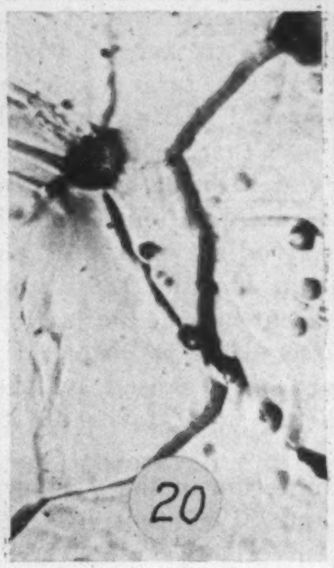


Fig. 20. Armco Iron. 900°C. Water Quench. 1000X.

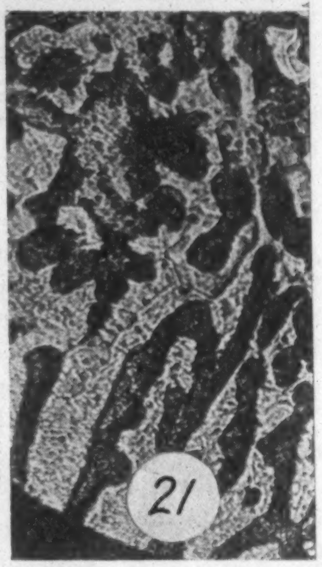


Fig. 21. Kroll's Fig. 11. 435X.

in that great variations in the *extent* of transformation are present. This was the usual case in both of the hand puddled specimens and is caused by the piling of muck bars of variable composition in the rerolling processes. The Aston product is rolled from a single large ball of spongy iron which gives a more uniform product. However, Rawdon and Knight of the U. S. Bureau of Standards[17] were not able to detect any great differences in physical or chemical properties in the 2 types of wrought iron, therefore the similarity shown in their reaction to heat treatment is to be expected.

The fifth material examined was a ¾" bar of Armco ingot iron. Figs. 19 and 20 show the structure upon quenching in water after 15 minutes at 900°C. In Fig. 19 at 100X an interlocking structure indicating at least 2 grain formations is apparent. In Fig. 20 at 1000X the "phantom," or γ network of Rawdon and Berglund[11] is shown. This figure shows the relief etching of the α structure and the shallow groove appearance of the γ network, both of which are characteristic of this specimen. Rawdon and Berglund[11] in their Figs. 29B, C, and D at high magnifications (1000X to 2500X) show the γ network to be made up of minute inclusions similar to those in Fig. 20 of the present paper. While their structures were taken from open hearth iron forged at high temperature and air cooled, Fig. 20 was obtained by heating a fully annealed α iron to just into the γ region and then quenching.

Before attempting to arrive at any conclusions regarding the structures here presented, 3 of the materials used were tested by thermal analysis. The differential method with straight differential plotting was used in preference to the derived differential method[18, 19] used by Burgess and Crowe[4] in their excellent determinations of the A_2 and A_3 points in pure iron. This was done at the sacrifice of sharper maxima at critical points in order to preserve certain changes in slope and gradual reversals which are masked by the derived differential method.

Individual Pt-Rh thermocouples were used in the differential and temperature circuits. A Leeds & Northrup horizontal coil electric furnace and a nickel neutral body and specimen holder were used. No constantly variable rheostat was available to control the heating and cooling rates so the method of constant heat input was used. A Leeds & Northrup potentiometer indicator permitted temperature readings to be taken to within ± 1.0°C. A moving coil, reflecting type of galvanometer placed at 40 inches from the ground glass scale gave differential deflections up to 70 mm. at the $\gamma \rightarrow \alpha$ transformation of Armco iron (Fig. 24). Each of the curves was checked at least once.

The rate of heating and cooling was about one-half as fast as for the representative curve for pure iron, Fig. 25, as obtained by Burgess and Crowe.[4] In 3 of the cooling curves the Ar_3 and Ar_2 points were determined by separate runs and the resulting curves connected together at the short horizontal lines on the graphs.

The table summarizes the results of thermal analysis.

DIFFERENTIAL THERMAL ANALYSIS RESULTS

	Weight of Sample Grams	Ac$_2$		Ac$_3$			Ar$_3$			Ar$_2$	
		Max.	Rate °/Sec. (742°-815°)	Begin	End	Rate °/Sec. (870°-985°)	Begin	End	Rate °/Sec. (985°-870°)	Max.	Rate °/Sec. (815°-742°)
Aston W. I. Fig. 22	25	769°	.051	897°	924°	.043	928°	902°	.062	769°	.064
1890 W. I. Fig. 23	19	769°	.057	(?)	924°	.044	932°	902°	.058	769°	.061
Armco Iron Fig. 24	20	769°	.084	915° Max.	——	.055	897°	890° Max.	.066	766°	.062
Elec. Iron (B. of Stds.) Fig. 25	32	770°	.130	918° Max.	——	.130	893°	870° Max.	.125	(?)	.125

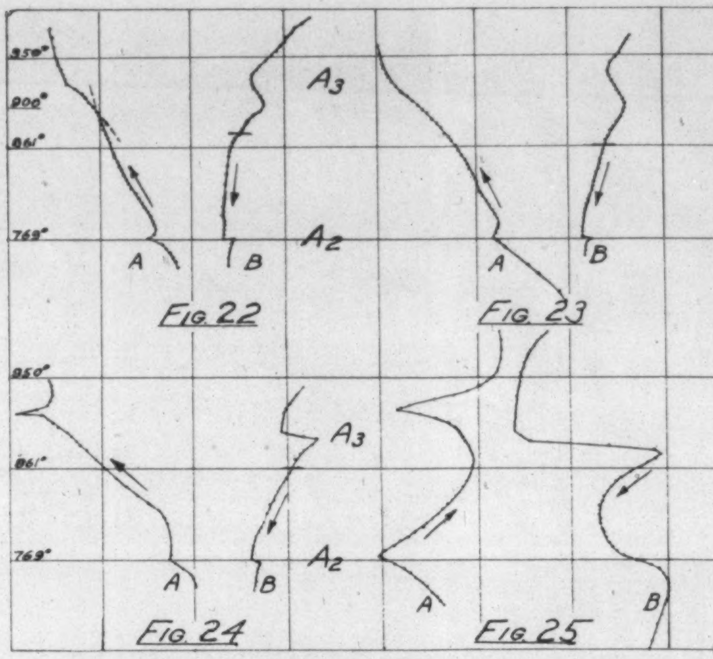
Fig. 25 was plotted from the data of Burgess and Crowe.[4] Although the maxima shift slightly from the curves as originally drawn using derived differential plotting, a good comparison may be drawn between Figs. 24 and 25. Considering the much greater sensitivity of their moving coil galvanometer, the larger specimen which they used, and their more rapid heating and cooling rates, the 2 sets of curves are in rather good agreement.

The wrought iron curves, Figs. 22 and 23, differ from the pure iron at the A_3 transformations. Upon heating, a gradual heat effect is observed between 897° and 924°C. for the Aston process iron. For the 1890 wrought iron this same reaction is apparent but the limits are not clear. Upon cooling, the reverse effect is observed and to a more marked extent, probably caused by the natural "drift" of the set-up being more favorable to cooling curves.

The elevation and elimination of the A_3 thermal critical points by high P and Si content in the work of Gough and Murphy[20] and others [21, 22, 23] indicates the general effect of these elements, an effect found to a lesser degree in the normal wrought irons dealt with in the present work because of their lower P and Si content. Rawdon and Knight[17] give 0.10% P, 0.12% Si in the analysis of Aston process wrought iron and 0.147% P, 0.22% Si in the analysis of the hand puddled iron used in their comparisons. The average for Aston muck bar was reported as 0.12% P and 0.142% Si by the same authors. This is in comparison with 0.48% P and 0.41% Si in the "defective" wrought iron of Gough and Murphy[20] in which the A_3 maxima were practically nil.

DISCUSSION

In reviewing the data here presented and that found in the literature, the work of Kroll[14] was found to have the most direct bearing on the special structures here presented. Fig. 21 is a reproduction of his Fig. 11, produced by etching for one hour in a weak picric acid solution. This material is a phosphoric iron containing 0.378% P and only 0.007% Si. Although the heat treatment was not stated the author infers that it was quenched in water. The relief network (compare with Fig. 10) is definitely associated with the P content by Kroll who states that the structure furnishes a rapid and trustworthy optical analysis of the phosphorus content. It appears from the present work that the quenching temperature would require very close control for successful operation, providing in the first place that P was the only element affecting the structure. The Aston wrought irons containing only about 1/3 the P content gave about the same amount of the relief network by quenching from 910°C. (See Fig. 6.)



Differential Thermal Analysis Curves
Ordinates: Temperature $-\Theta_{Fe}$
Abscissas: Differential Deflection $-(\Theta_{Fe}-\Theta_{Ni})$
1/70 of actual deflection in Fig. 25.
1/20 of actual deflection in all others.

According to Kroll the isolated patches are rich in P and represent pseudomorphs of individual γ iron grains. Kroll also states that the α ferrite occurs at a lower temperature in the parts rich in P. The more recent work of Haughton[23] on the alloys of iron and phosphorus proves quite conclusively that P raises the A_3 transformation very rapidly, therefore ferrite would form at lower temperatures in the parts *poor* in P (on cooling). Haughton's Fig. 8, reproduced here as Fig. 26 shows that the γ phase is completely eliminated at about 0.55% P, the temperature at which this maximum amount of P may be present in γ iron being about 1040°C.

Silicon has also been shown to have this same influence in limiting the range of existence of γ iron. This work is summarized in Haughton and Becker's diagram, Fig. 1.[22] The effect of Si is not as drastic, about 2% being the maximum amount in the γ iron loop.

The following explanation has been considered by the writer on the basis of the microscopic and thermal data presented, the comparatively high P and Si contents of ordinary wrought iron as compared with pure iron, and the diffusion effects of P.

The P present (about 0.12% total) is subject to segregation. Its relatively slow movement in iron is well known. Approximately one-half of the P content of wrought iron has been found to be present in the slag,[17] and it will first be assumed that the areas directly surrounding the slag have a relatively low P content. Upon heating a specimen, say to 910°C., these P-poor areas will transform to γ iron while the remainder of the ferrite will not (See Fig. 26). According to the work of Whiteley:[24] "When ordinary hypoeutectoid steel is heated to, and maintained at, temperatures between the critical points Ac_1 and Ac_3, part of the phosphorus contained in the γ iron diffuses into the adjacent ferrite owing to the fact that, under these conditions, the phosphorus is more soluble in the latter than in the former." If this condition still applies to very low carbon irons, the newly formed γ grains at 910°C. will lose their P to the remaining α ferrite thus further impoverishing them in P. On rapid cooling the γ grains will transform to α, taking the crystalline orientation of the adjacent α grains (See Fig. 11) but maintaining this low P content. Upon etching with such reagents as picric acid, the relative purity of the former γ regions causes them to remain in relief from the surrounding structure. Upon further etching with a cupric reagent, copper is deposited upon the former γ regions, giving evidence of their lower P content (See Figs. 12 and 17). Upon slow cooling from 910°C., however, the γ regions after transforming to α

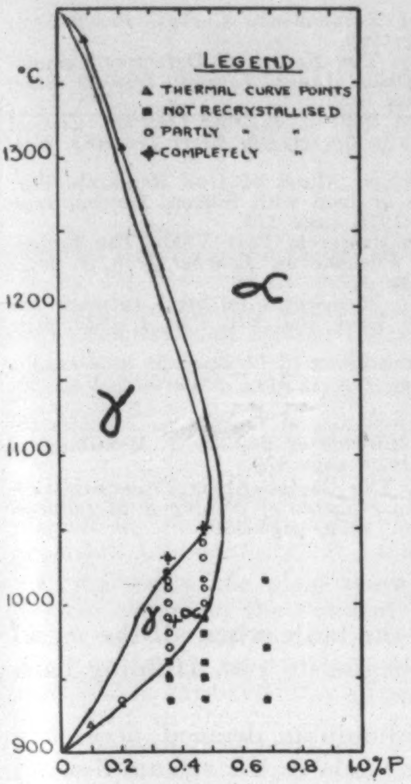


Fig. 26. The Range of the $\alpha + \gamma$. Region (Haughton).

have ample time to receive P by migration from their relatively P-rich surroundings. In fact a rapid quench is required to maintain at room temperature the full effect of the γ markings (compare Figs. 4 and 5). This follows from Whiteley's[24] observation: "At temperatures below 650°C., phosphorus diffuses extremely slowly in ferrite, above this temperature, the velocity gradually increases, and above 800°C. it is very rapid, for a diffusion which requires 4 hours to complete at 740°C. is complete in 15 minutes at 830°C." Thus quenching is necessary to prevent some diffusion in the neighborhood of 900°C.

A qualitative test was made for the rate of diffusion of P in the material at hand. The specimen quenched in water from 910° was reheated to 880°C. in one hour and again quenched. Upon etching and observation the special areas were no longer visible, a slight haziness of the grain boundaries immediately surrounding the large slag patches being the only remnant of the special structure—the P content of these areas having been increased by diffusion from the surrounding ferrite.

It follows that heating to higher or lower temperatures than 910°C. merely causes more or less of the iron to transform to γ, lose P, and then recrystallize to a P-poor α upon quenching. Other impurities in the iron, such as oxygen, may well exert a similar but as yet undetermined effect in the $\gamma \rightarrow \alpha$ transformation. Si quite definitely is in this class although its rate of diffusion is faster.

If the fundamental assumption that the regions surrounding slag patches are low in P is incorrect, some other forces must cause the $\alpha \rightarrow \gamma$ transformation to begin in these areas (See Figs. 3 and 4). In any case the final condition of low P in these areas is quite definite as shown by the phosphorus indicating etch (Figs. 12 and 17).

It is believed that the thermal analysis data bear out these conclusions in that the $\gamma - \alpha$ transformation region is both raised and broadened in temperature over that found in pure iron by the same methods.

It would be most interesting to view microscopically the transformation of a polished and etched wrought iron section in the hydrogen furnace developed by B. A. Rogers.[26] It is the writer's opinion that the transformation could be seen to progress in an orderly manner from slag particles and grain boundaries as indicated by the photomicrographs. The vacuum heating, mercury quenching procedure used by Albert Sauveur and C. H. Chou[27] might also yield new or confirmatory information if applied to wrought iron.

The inclusion theory of the γ network[11] has been further substantiated here in observations of wrought iron and pure iron. While it is not contended that all of the γ grain boundaries are very high in inclusions, it is believed that there is a definite tendency for the impurities to migrate to the boundaries, giving evidence of the γ network after heating but slightly into the γ region and then quenching (Fig. 20).

The α veining observations are in accordance with others previously published excepting with those claiming that each large ferrite grain has its own veining network, independent of the network within the large adjacent grains. The veining network in its fully developed state was here observed to be usually independent of the large grain boundaries (See Figs. 14 and 15).

The strengthening of the α-veining network by slow cooling from 1000°C., Fig. 14, as compared with that produced by quenching conforms with either of the 2 theories of veining formation. The slow cooling may allow the submicroscopic inclusions to take positions in the veining boundaries during their formation, in accordance with the theory of Bannister and Jones.[12] Slow cooling should also be favorable to the growth of a relatively few α nuclei in each γ iron grain, giving fairly well defined veining boundaries as in Storey's theory.[1]

No fully satisfactory explanation of the formation of structures as in Figs. 14 and 15 may be drawn from the present work. The most certain statement that can be made concerning α-veining formation is that it occurs at the $\gamma \rightarrow \alpha$ transformation.

The practical effect of P in wrought iron has been determined quite recently by Higgins[25] who is in agreement with earlier investigators in finding that P lowers ductility and increases the strength. The beneficial effect of P on the endurance limit, as reported by Higgins, is not as widely accepted, due perhaps to natural prejudice against P in steel.

It is conceivable that under proper control, quenching treatments such as reported here might be useful in controlling phosphorus in the production of wrought iron.

CONCLUSIONS

1. The temperature range of the $\gamma - \alpha$ transformation in wrought iron is broader than that for purer iron.
2. The relatively high content of P and other impurities in wrought iron permits certain diffusions to take place in the temperature region of the transformation.
3. As a consequence to 1 and 2, the positions of the α and γ grain markings present during transformation may be observed at room temperature if the wrought iron is quenched in water from the transformation region.
4. The γ "phantom" network was confirmed in pure iron and observed in wrought iron, and was found to contain visible inclusions.
5. Alpha-veining was observed in Aston wrought iron, particularly after slow cooling from above the transformation temperature range.
6. Aston process and hand puddled wrought iron have the same characteristics at the $\gamma - \alpha$ transformation.

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Electric Steel for Castings

(Continued from page 219)

made steel does not need an aluminum addition to deoxidize it, the aluminum being added to the steel for castings to prevent the porous condition arising as the result of the injection of mold gases. With green sand molds of proper constitution a large aluminum addition is unnecessary.

Therefore it seems that if steel free from this sulphide formation and resultant depreciated ductility is to be produced, it is necessary to:

1. Produce the steel so that it contains an amount of iron oxide equilibric with the carbon and manganese contents.
2. Limit the presence of silicon to about 0.25% so that it may exert the minimum depreciating effect upon the iron oxide equilibrium.
3. Make the molds so that the minimum of aluminum content is necessary.
4. Or lower the sulphur content so that no "envelopes" can be found.

With the more advanced knowledge available today regarding sand control and the effect of variations in mold permeability, strength and moisture content, there exists no reason why green sand molds cannot be produced which require the minimum of aluminum addition to the steel to produce sound castings.

It does not seem improbable that a material for use in the acid furnace with a sulphur content appreciably lower than at present, will soon be available.

The pure alumina or aluminum silicate inclusions usually found in aluminum treated steels also play a large part in depreciating the ductility of the metal and every effort should be made to prevent their formation.

Remembering the ability of aluminum to reduce the oxides of silicon, manganese and iron it seems advisable that extreme care should be taken in making the addition of this element to the steel. If it is thrown in the stream during the tapping operation it is quite liable to become oxidized either through reacting with any of the above oxides if dirty steel is being tapped or if the slag is allowed to run with the metal or possibly by reaction with atmospheric oxygen. Therefore it is suggested that the aluminum be added by wiring it to a heavy rod and plunging it to the bottom of the ladle when all the metal is in the ladle and the slag begins to run. Holding back the slag is essential.

Any further addition of aluminum deemed necessary in the hand shanks should be made in the stream flowing from the large ladle so as to avoid atmospheric oxidation of aluminum at this time.

Conclusion

As stated previously, while the theories advanced in this paper may be in error, their application has resulted in eliminating several of the great variations connected with acid electric steel practice and in producing better steel castings. They are by no means a panacea. From a purely metallurgical view point, however, it may be said that better steel is bound to result from any such methods as discussed, entailing as they do an increased knowledge of slag significance. It is hoped that in reading this, even though some statements may be disputed, further thought along these lines will be stimulated and it is hoped that this thought may aid in someday producing a material for the steel founder with greater fluidity at normal casting temperatures, the ability to better withstand the damaging influences of mold gases, and the property of higher strength near solidification temperature.

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Percent Elongation in Tensile Test

(Continued from page 225)

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READERS' COMMENTS

The Precipitation Theory of Steel Hardness

By ROY M. ALLEN

THE precipitation theory of the hardening of steel proposed by Dr. Sauveur in the August issue of METALS & ALLOYS, supplemented by his expressed hope of discussion, prompts the presentation of further theorization along the same line. For some time past the writer has held his own idea of the mechanism of steel hardening, but its radical nature has been responsible for keeping it in abeyance, at least in so far as publication has been concerned.

Dr. Sauveur's precipitation theory comes so close to coinciding with it, however, that the additional radical element present in the writer's theory may not cause too much of an explosion in metallurgical ranks if offered at this time.

Undoubtedly the moot question as to whether the carbon of steel occurs in the form of Fe_3C molecules in solution in γ-iron, or as plain carbon atoms only, will never be settled by direct evidence, yet we must start with it in discussing all the possibilities involved in a precipitation theory. If the carbon is present as the carbide (Fe_3C) then obviously it must be precipitated out as such and Dr. Sauveur's explanation would seem to fit all the observed facts. If on the other hand, as seems more plausible, the carbon is dispersed throughout the γ-iron as a true solid solution of carbon, then another condition must be satisfied if the mechanism of precipitation is to be exactly in accordance with that suggested by Dr. Sauveur. That is, when the $\gamma \rightarrow \alpha$ transformation occurs, each atom of carbon must first combine with 3 atoms of iron and form the crystallographic arrangement of the carbide molecule, involving an additional time element; otherwise at the outset pure carbon particles would be precipitated, not carbide particles.

On first thought, this alternative precipitation condition (i.e., of pure carbon) seems too radical to fit in with our preconceived ideas of the relationship existing between iron and carbon, but possibly the true nature of hardened steel is still a subject of controversy as it has not received due consideration.

The fact that in steel we have, not merely one element subject to allotropic changes, but two, also seems to have been overlooked in trying to formulate a satisfactory theory.

Let us consider some of the possibilities involved, on the assumption that pure carbon, as well as carbide, may be a factor in the precipitation theory of the hardening of steel.

Carbon occurs in at least 3 allotropic forms: (a) as the diamond, in the isometric system; (b) as graphite (Rhombohedral); and (c) amorphous (probably). Of the three, the diamond form is the hardest substance known; therefore if we admit the possibility of carbon being precipitated as such, under conditions where the time element is too short to allow the formation of all the carbon present as carbide, it is easily

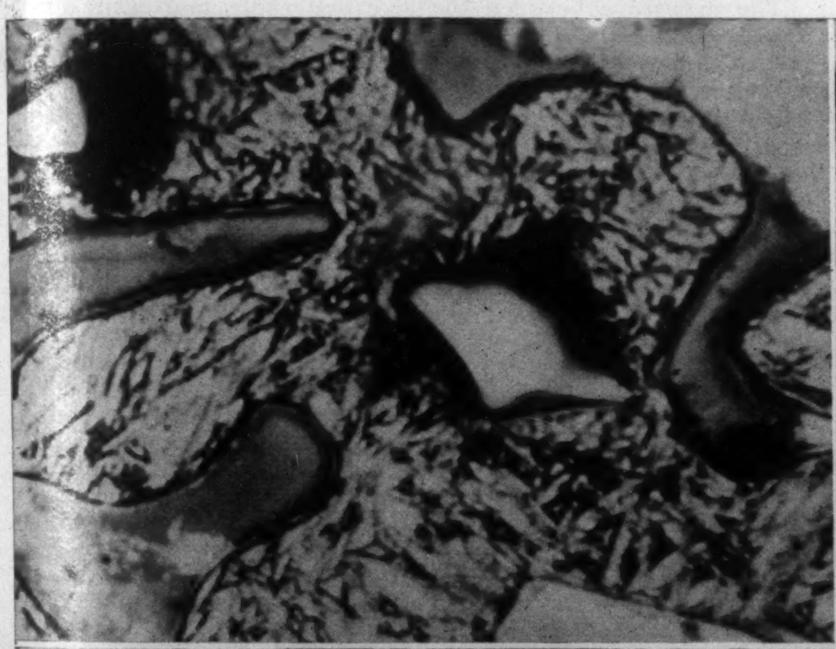

Fig. 1. Diamond crystals (2), in martensitic chilled cast iron. Magnified 3000X. The larger shows the octahedral form while the other gives a good idea of the dark constituent (graphite) always associated with the diamonds.

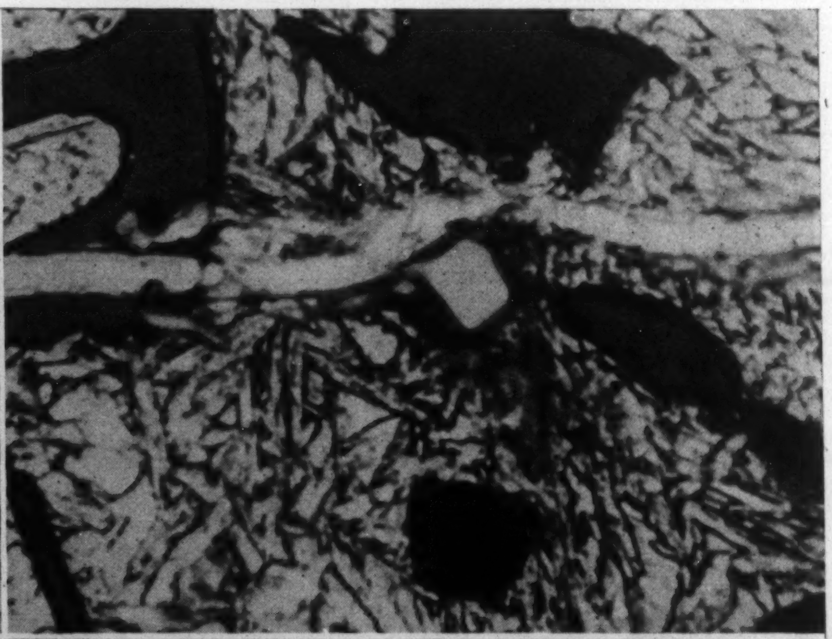

Fig. 2. Microcharacter scratch passing a diamond crystal (from left to right) in martensitic chilled cast iron. Magnified 3000X. The scratch is deflected around the crystal due to play in the instrument.

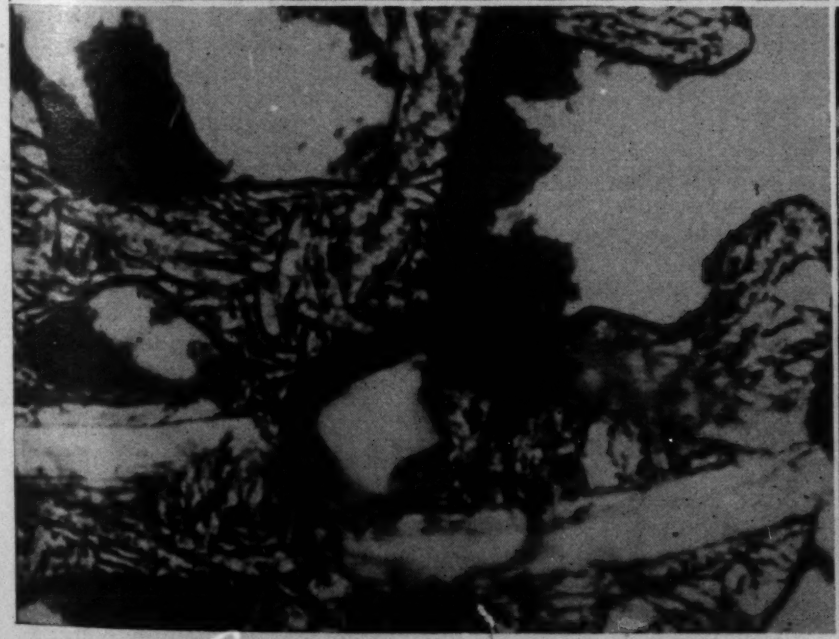

Fig. 3. Microcharacter scratch passing a diamond crystal (from left to right) in martensitic chilled cast iron. Magnified 3000X In this case the instrument has jumped after coming in contact with the crystal and has resumed motion at the side.

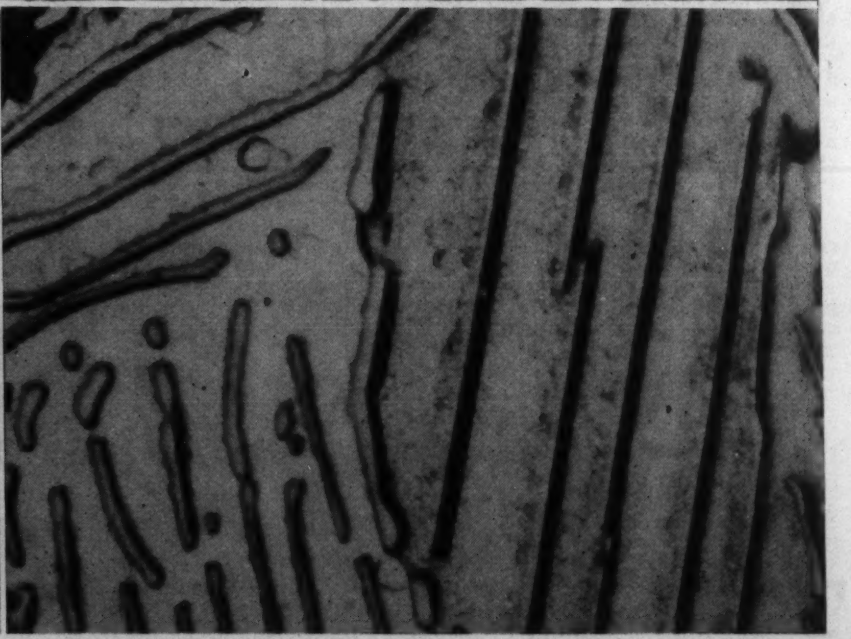

Fig. 4. Cementite plates in pearlite lying in planes of maximum weakness of the alpha iron grains. Portions of at least four grains are included in the area shown.

conceivable that several atoms could cluster together as a nucleus, and under the tremenduous pressure involved, become crystallographically oriented as submicroscopical diamonds.

That such is not only possible, but actually does occur under some similar conditions, has been demonstrated by the writer. Some years ago, while carrying on a series of experiments with martensitic chilled iron, microscopic diamonds* were found to be present in some of the chills, and subsequent examinations of chilled irons have demonstrated that they are far more common than originally thought.

Fig. 1 shows the appearance of such a diamond crystal at a magnification of 3000 diameters. The characteristic octahedral shape, clarity and high refractive index are clearly shown.** Without going into detail as to all the proofs of the nature of these inclusions, it may suffice to show a couple of views of the reaction they show to a microcharacter scratch. Fig. 2 shows that the diamond of the Microcharacter will not pass across these crystals but is forced to go around them (due to play in the instrument) while Fig. 3 shows what happens when the scratching diamond hits dead on. The point has jumped entirely off the surface and started a new line at the side!

If diamond crystals of *microscopic* size (up to 10 micra, axis diameter) can form in suddenly cooled cast iron, it would seem a simple matter for *sub*-microscopic carbon crystals, corresponding to the diamond crystallographically, to be precipitated out of any iron-carbon mixture (especially when rich in carbon) with rapid cooling through the transformation range. However, it would probably be true that some of the carbon would succeed in combining with iron to form the carbide, so that the latter would also be present.

Therefore if such a condition does occur, we would merely modify Dr. Sauveur's definition of martensite as follows: "An aggregation of α-iron with carbide particles and crystallized carbon (as diamond) of submicroscopic size." Certainly such would account for any degree of hardness present in martensite.

We must then consider what takes place when martensite decomposes to troostite. It is at this point that the allotropic property of carbon may play an important part, for it seems likely that carbon when crystallized in the closely packed form of the diamond, is not directly soluble in either γ- or α-iron, but must first transform to graphite or the amorphous state.

We do not know under what conditions carbon changes over from one allotropic form to another, but judging from the rare occurrence of the diamond in nature and the abundance of graphite, not only massive but as microscopically dispersed particles in calcite, limestone, schists, etc., where conditions should at least occasionally be conducive to diamond formation, it would seem that there must be a marked tendency for diamond carbon to change over to graphite or the amorphous state at elevated temperatures when subjected to strain. This line of reasoning also seems to be borne out by the appearance of a dark constituent generally associated with the diamond crystals in chilled iron (Figs. 1, 2, and 3), which is undoubtedly graphite.

It would thus seem that troostite may represent a transitory state of extremely narrow range where the diamond carbon is changing over to graphite or amorphous carbon preparatory to combining with iron to form the carbide, this latter reaction proceeding simultaneously of course, with the decomposition of the diamond carbon so that at any moment, up to the time when the process is complete, at least 2 of the possible 3 forms of carbon would be present, i.e., diamond carbon, graphite (or amorphous) and carbide. The former would be the first to disappear and the latter the last to appear; when the reaction is complete only the carbide would be present. That such a condition is possible is apparent when the mobility of carbon atoms with respect to iron atoms at the temperatures involved, is considered. The carbon in the free state might range from atomic through submicroscopic to microscopic, but in any case would readily account for the deeply colored appearance of troostite. Such explanation of the apparent intensive etching of this constituent seems more simple of comprehension than that suggested by Dr. Sauveur.

The difference between the so-called white martensite and dark martensite can be explained in a similar manner, as representing merely a minimum (or zero) amount of amorphous carbon in the former and correspondingly increasing amounts in the latter. According to this idea, white martensite should be somewhat harder than the darker kinds.

On the basis of this precipitation theory, the proposed definition of troostite would be modified as follows: "An aggregate of α-iron with carbide and amorphous (or graphitic) carbon particles of submicroscopic to microscopic size."

Given sufficient time for all the carbon to become associated with iron in proper crystallographic arrangement, the result would be sorbite, and the description of this constituent and of pearlite as proposed by Dr. Sauveur would be accurate. The essential difference between sorbite and pearlite in this case lies in the fact that each individual grain of α-iron—with its included cementite—attempts to reach a condition of maximum stability by forcing the cementite to occupy positions along the planes of greatest weakness (i.e., slip or cleavage planes). Sorbite represents the extreme condition of nonstability and pearlite the end point in maximum stability. The arrangement of the cementite of each grain along crystallographic planes in the stable condition is well shown in Fig. 4, in which parts of several grains are visible.

This theory fits in with the view that on slow cooling austenite decomposes directly to pearlite, for with ample time factor, each carbon atom would combine directly with 3 of iron and free carbon would not be precipitated at all, in any form. In this connection, however, it would seem to simplify matters to conceive of pearlite as differing structurally from a true eutectic only in its method of forming and that the eutectic composition (of α-iron and cementite) occurs entirely within the confines of individual grains instead of in a massive condition.

It is hoped that this radical modification of Dr. Sauveur's theory will not be rejected by an ex cathedra metallurgical mind without first giving it an unbiased thorough consideration.

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*As to *mode of occurrence,* essentials for producing diamonds were present, i.e., free carbon, high temperature and pressure.

Crystallographically they have all the earmarks of the diamond, isometric system, typical habit the octahedron, with frequently perfect 111 faces but many others can be identified. Percentage of abnormality in crystallographic development about the same as in natural diamonds and the nature of the abnormalities identical.

Optically the crystals are isotropic, refractive index, as determined by a modified Beck method, corresponding to the diamond, dispersion obviously high (but cannot be easily measured), luster adamantine, colorless, transparent. As a matter of fact, the refractive index and adamantine luster alone establish their positive identity to a mineralogist, even if not a single other characteristic could be determined, as no other known mineral or element possesses similar properties.

When we come to the question of *physical properties,* obviously the only one that can be determined is hardness, but in this respect also, the diamond stands alone, so that micrographs 2 and 3 tell the full story. While rated at 10 in the Mohs scale of hardness, the sapphire being 9 and the nearest competitors of the diamond, carbides of silicon, chromium, tungsten, etc., some metallic nitrides and possibly boron, being somewhat in excess of 9, the actual difference between the diamond and any other known substance is still enormous.

Another condition which demonstrates the hardness of the inclusions is that neither the carborundum polishing papers or alumina used in preparing the surface of the metal, have any effect whatsoever on them, all corners and angles which project above the plane of the finished specimen, remaining perfectly sharp. The same polishing papers and alumina will prepare carbides and nitrided surfaces perfectly flat and true. The appearance of the crystals on the polished surface of the metal must be seen with a stereoscopic binocular eyepiece to be appreciated. They look very much like a glass cutting diamond in its setting.

Chemically they are absolutely inert to any of the etching reagents and do not respond as carbides to special reagents which serve to demonstrate the latter satisfactorily.

**The etching solution employed in this case was boiling sodium picrate, while the plane of the focus is a compromise so as to bring out the diamond, which has polished is relief.

On The Formation of Lamellar Pearlite

By R. G. GUTHRIE and J. A. COMSTOCK

THE VERY admirable article by Dr. Albert Sauveur, appearing in August issue of METALS AND ALLOYS under the title of "Steel in the Light of the Precipitation Theory" undoubtedly has caused considerable serious thought on this most important subject.

The present writers read with satisfaction the development of the precipitation theory as applied to the quenching of steels. However, we are not convinced that the complete transformation:

austenite → martensite → troostite → sorbite → pearlite must invariably take place on quenching and tempering, except where typical acicular martensite is produced at room temperature on quenching.

As much to keep the discussion open, as to state our own beliefs on the subject, we submit the following:

It has been suggested* that the Ar_1 transformation may be a two-part reaction:

(1) Allotropic change from gamma to alpha iron.
(2) Formation and precipitation of iron carbide or
austenite → ferrite + cementite.

Is it not logical to contend that the formation and nature of carbide particles precipitated in the ferrite on cooling through the transformation is dependent on the temperature of the transformation? That is to say, a drastic quench will retard the allotropic transformation to around 300°C., at which temperature the new-born ferrite is too rigid to favor diffusion and sub-microscopic carbide particles are precipitated. We recognize the resultant hard microconstituent as acicular martensite. If the quench is less drastic so that the allotropic transformation takes place at a certain higher temperature, then the precipitated carbide particles are larger and we recognize the microconstituent as primary or nodular troostite. A still higher temperature of transformation produces primary sorbite, and the highest temperature at which the transformation will take place produces primary or lamellar pearlite.

We believe that tempering acicular martensite at successively higher temperatures produces, by coalescence of carbide particles, the microconstituents, respectively:

martensite → secondary (granular) troostite → secondary sorbite → secondary (granular) pearlite

or: tempering primary troostite would produce successively secondary sorbite and secondary pearlite;

or: tempering primary sorbite would produce secondary pearlite.

To illustrate the contention that acicular martensite is not necessarily the progenitor of pearlite, and, we believe, is never the progenitor of either primary troostite, primary sorbite, or primary pearlite, the simple experiment was performed whereby austenite and lamellar pearlite were produced in equilibrium with each other.

Using a thin strip of eutectoid composition, plain carbon steel, the strip was first heated throughout at 1800°F. to austenize the steel, and then transferred to a lead pot and partly immersed in lead at about 1300°F. After two minutes in the lead, the strip was quenched in water.

Since the critical temperature on cooling for this steel is approximately 1290°F., obviously some part of the strip was quenched at just the correct time and temperature to have, prior to quenching, austenite on the hot side and decomposed austenite on the cool side. Actually, there is a gradation or critical zone where austenite and decomposed austenite are in equilibrium and co-exist much as ice and water can co-exist at 32°F.

On quenching this critical zone, the austenite transforms to martensite and the decomposed austenite remains *status quo*.

The accompanying photomicrograph, taken with a 2 mm. 1.40 N.A. oil immersion apochromatic lens at a magnification of 4000, shows the co-existence of martensite and lamellar pearlite with no intervening secondary constituents.

In the case of formation of all the primary microconstituents excepting lamellar pearlite, the effect of super-cooling enters into the mechanism of the transformation in varying degrees. Consequently, the relationship existing between the allotropic change and the carbide formation and precipitation in these cases may not be identical with said relationship in the case of the formation of lamellar pearlite. In the latter instance, the carbide is formed and precipitated, you might say, just as the cards are being shuffled (recrystallization) and with the freedom corresponding to the maximum temperature. As a result, cementite is rejected to the preferred planes of crystallization in the forming ferrite to produce a typical laminated structure.

Where conditions of super-cooling prevail, as in the formation of primary microconstituents martensite, troostite, and sorbite, the carbide is precipitated and retained almost in situ in the ferrite as an aggregate.

The writers trust that their effort to contribute to the invited discussion of the first-mentioned article will be received with the same feeling of collaboration with which it is written.

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*Kotaro Honda: "On the Formation of Martensite in Carbon Steels."

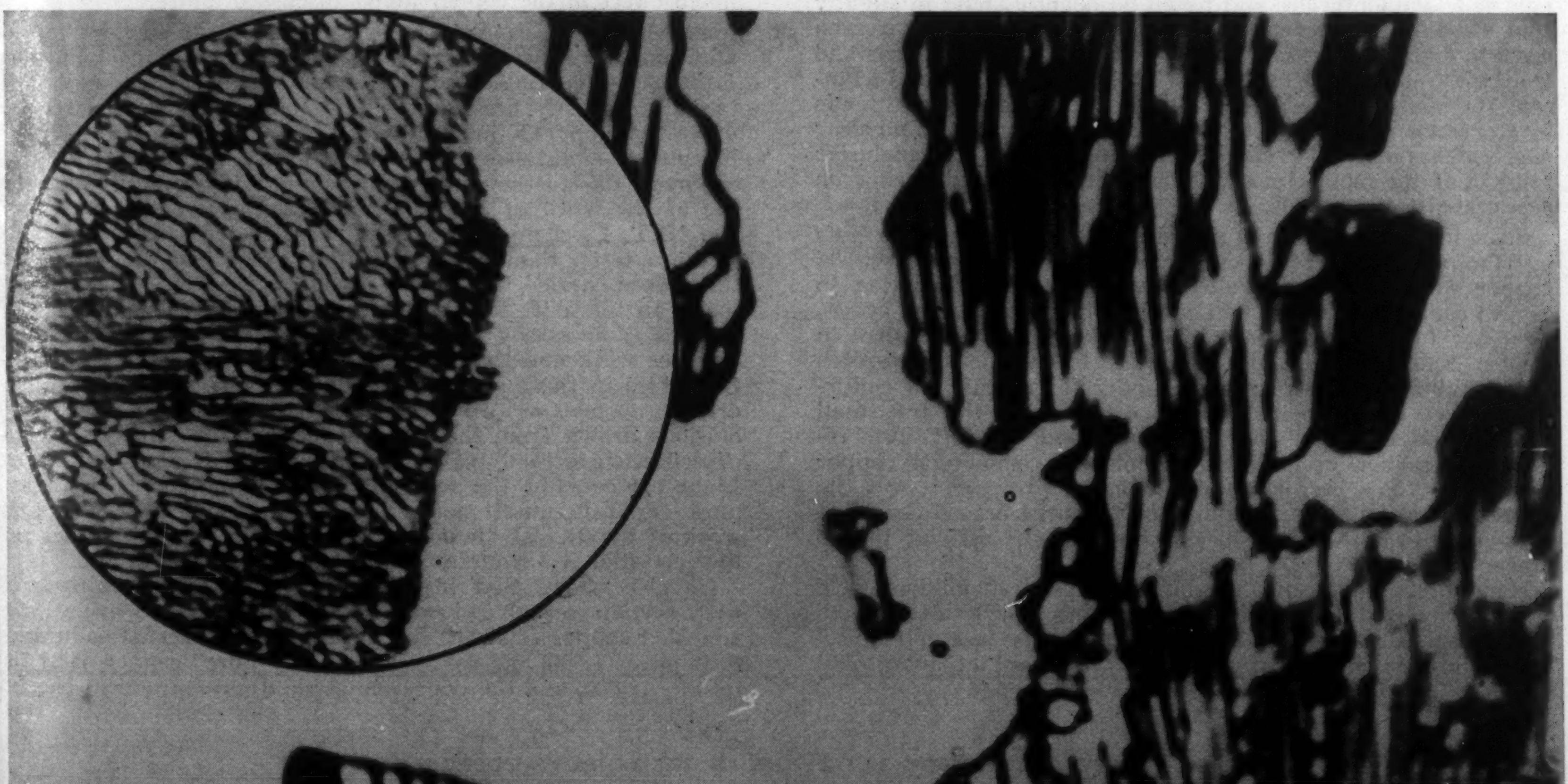

Fig. 1. Shows martensite and lamellar pearlite co-existing without intervening secondary constituents. Further etching with nital develops the acicular structure of the martensite but over-etches the pearlite. Mag. 4000X.

Fig. 2. (in insert) Shows martensite and lamellar pearlite co-existing without intervening secondary constituents. Mag. 2500 X.

The Resistance of Copper and Its Alloys to Repeated Stress

A Correlated Abstract by H. W. Gillett†

Part II. Effect of Impurities in and Additions to Copper. Lack of Satisfactory Accelerated Test Methods.

THERE ARE no data on the endurance of pure copper. All published data are on tough pitch copper, which is a copper-copper oxide alloy. There has been recently put on the market a so-called "high conductivity oxygen-free copper[1] in which oxygen is absent or at a minimum. While the iron content is probably a trifle above that of ordinary refined copper, since the conductivity is given as a trifle below that of tough pitch copper, its total impurity content is low. This material is described as showing about the same tensile strength as tough-pitch copper, but as having a somewhat higher ductility. Tests are cited to show that in bend and twist tests it is superior to tough pitch, and it is stated that it offers "more resistance to fatigue than ordinary electrolytic copper," but no actual endurance tests are given. It is stated[2] that in tests in progress in comparison with tough-pitch (static properties, degree of cold working of the specimens and number of cycles of test not stated), it gives about 17,000 lbs./in.^{2}, while tough-pitch gives 15,000 lbs./in^2.

This is the only information available on oxygen-free copper not contaminated by some excess deoxidizer; so there is no true base-line for comparison of the effect of various additions and impurities. Nor are there any actual endurance data for annealed copper containing various amounts of copper oxide. Gillett and Mack[3] report some preliminary work carried out only to 50 million cycles and found no definite trend in slightly cold worked (final reduction 16.3%) copper of varying oxygen content, or that with small amounts of residual phosphorus and manganese used as deoxidizers (Table 1). Gough[4] and Hanson, Marryat and Ford[5] report so-called endurance tests made by an "accelerated test method," which likewise showed no clear effect of oxygen, but which cannot be accepted as conclusive because of the method used. Comment will be made below on accelerated test methods.

Some indirect evidence is afforded by actual endurance tests by Hanson and Marryat[6] on specimens of annealed arsenical copper with about the same arsenic and some variation in oxygen content, which, though carried only to 20 million cycles, (too few for finality) showed no marked effect of change in oxygen content within or somewhat below the tough-pitch range, although the strengthening effect of arsenic, on endurance as well as on tensile strength, especially of the first small additions, is quite definite. Memmler and Laute[7]* also reported 20 million cycle tests on annealed commercial copper with about 0.25% As, which show some improvement over the figures for arsenic-free copper, but the presence of considerable amounts of Ni, Sn, Sb, and Pb makes it difficult to appraise the results.

Hanson, Archbutt and Ford[8] also report 20 million cycle tests on annealed copper-phosphorus alloys which clearly show that phosphorus improves the endurance resistance of copper. The endurance ratio is approximately constant at about 0.40 with 0.03% to 1% P.

Data from Hanson and Ford[9] on copper-bismuth alloys very low in oxygen, being available both on annealed and cold worked material, has already been given.** The results indicate that such amounts of bismuth as can be tolerated in copper on the basis of static and impact properties have no especial effect on endurance. Archbutt and Prytherch[10] studied a series of annealed copper-antimony and copper-arsenic-antimony alloys (which were lower in oxygen than ordinary tough pitch copper), which showed that antimony has much the same effect in improving endurance as arsenic does, and verified work of Hanson and Marryat on the effect of arsenic. Of the three elements As, Sb, and P, of which rather small amounts (0.10 to 0.25% As, 0.20% Sb, 0.15% P) will raise the 20 million cycle endurance of copper to about 15,000 lbs./in.^{2}, P in that amount injures the electrical conductivity the most. The small amount of residual phosphorus in commercial phosphorized copper does not appear to affect the endurance appreciably.

Johnson[11] states that the resistance to fatigue of copper alloyed with small amounts of nickel is greater than that of arsenical copper, but cites no references.

Other comment on the effect of arsenic as cited by Bamford[12] has already been given.***

There are no data on corrosion fatigue or in regard to the notch effect in the copper alloys discussed above.

Accelerated Tests. Practically all of the data on the effect of additions to copper are from the work of the National Physical Laboratory in England, which work is, in most details, extremely carefully done and reliable. Unfortunately the work, up to that reported in 1927, is not of the same degree of credibility as to endurance properties as it is to the other properties reported, due to the use of an "accelerated" test. In the middle of the work of Hanson and Marryat,[6] reported in 1927, the N. P. L. changed from the "accelerated" test to real endurance tests, though unfortunately even then the endurance tests were carried only to 20 million cycles. No comment was made on the shift in methods, so that careful reading of the articles is necessary to differentiate between the credible endurance work and that which it is not certain is correct, since the tables in those articles include data for "endurance properties," no matter by which method they were obtained. Conclusions drawn from the "load-deflection," or "change-in-rate-of-deflection method" of these "accelerated" tests may happen to be correct, but the reliability of that method is nowadays most seriously questioned, and its abandonment in the later work of the N. P. L. indicates that the N. P. L. itself does not now consider it adequate for the securing of valid data.

Gough[13] shows that his deflection tests agreed fairly well with certain actual endurance tests to 30 million cycles; for annealed copper of 32,500 lbs./in.^{2} tensile, his actual endurance tests gave 11,000, and for "commercially pure" copper of 29,500 tensile, 12,500, and the deflection method gave practical checks.

TABLE 1A. EFFECT OF OXYGEN IN COLD-DRAWN COPPER

								STATIC			ENDURANCE		
No.	Worker	Mark	Composition				Anneal	Tensile Strength lbs./in.^{2}	Elong. %	Red. Area %	Stress, 1 Million Cycles lbs./in.^{2}	50 Million Cycles lbs./in.^{2}	Endurance Ratio
	Effect of Oxygen		*Deoxidized Cu*	*P*	*Fe*	*Mn*							
1	Gillett & Mack	5	99.92	.01	.06	.02	Cold-drawn	36,000	32	84	20,500	17,500	.49
			Cu	*O*									
2		4	99.95	.036			Cold-drawn	37,500	30	73	21,000	18,500	.47
3		3	99.94	.049			Cold-drawn	37,500	29	67½	21,000	17,500	.46
4		2	99.90	.094			Cold-drawn	38,000	27	63	21,500	19,000	.50
5		1	99.75	.22			Cold-drawn	41,000	26½	48½	23,000	17,000	.41

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†Battelle Memorial Institute.

*See Table 3a, Part I of this summary—*Metals & Alloys,* Vol. 3, Sept. 1932, page 203.

**See Table 3b, Part I of this summary. *Metals & Alloys,* Vol. 3, Sept. 1932, page 203.

***Part I of this summary. *Metals & Alloys,* Vol. 3, Sept. 1932, page 204.

(The discrepancy between the endurance ratios of the two materials is not explained.) He admits later in his book, however, that the method is not applicable to non-ferrous metals and alloys. One can indeed see from other of Gough's[14] own data that the deflection method may not always give true results. He found that by stressing annealed copper at 16,000 lbs./in.^{2} for 350,000 cycles, thus damaging it by over-stressing, the deflection method showed the same indicated endurance limit (12,500 lbs./in.^{2}) as it did on virgin material, but the over-stressed specimen, loaded again, and only to 11,500 lbs./in.^{2}, broke in less than a million cycles.

Lessells[15] cites a case where the load-deflection method, applied to cold-worked copper, indicated an endurance limit of about 24,500 lbs./in.^{2}, while the material, loaded to 20,000 lbs./in^2, broke after a million cycles.

The A.S.T.M. Research Committee on Fatigue of Metals[16] says, "The rise of temperature and change in rate of deflection methods are only fairly satisfactory on wrought ferrous metals and fail to indicate correctly the endurance limit of a number of non-ferrous metals."

We are, therefore, forced to put the N. P. L. accelerated test data in a separate category and label it "not proven." It is quite probable that the conclusions indicated by the N. P. L. data, collected in Table 2 on effect of oxygen from the work of Gough[15] and Hanson, Marryat and Ford,[5] of iron, from that of Hanson and Ford,[17] and that on arsenic alone from that of Hanson and Marryat,[6] i. e., that normal amounts of the first two have no notable effect on endurance, and (as is indicated by other evidence) that arsenic improves the endurance properties, may be quite correct, but the accelerated test values given are not valid proof. Unless the endurance ratio changes markedly with increasing oxygen, one would, from the static properties, expect increased endurance with oxygen increased above the usual tough pitch limit, but little effect below it, while the available data indicate that the latter is probably true, while the former probably is not, so that very high oxygen probably means lower endurance than that for tough pitch.

The accelerated test method indicates that normal amounts of iron in copper would have no appreciable effect on endurance but that very large amounts (around 2%) would give an alloy of high endurance. These points deserve more definite proof, and it is to be hoped that the N. P. L. will check them on its reserve stock of material by actual endurance tests.

There is no easy road to the determination of true endurance data, especially on non-ferrous materials.

The only accelerated endurance tests yet suggested that appear to have any precise applicability to copper alloys are the accelerated torsion test of McAdam[18] and the Ikeda electrical resistance test.[19] The latter shows fair agreement with true endurance tests on the few non-ferrous alloys so far tested, but it would be rushing matters to accept it at present. It deserves further study.

No categorical statement as to the number of cycles to which an endurance test must be run to give valid data can be made for non-ferrous alloys. There are few steels, and these are chiefly those of extraordinarily high tensile strength, on which it is necessary for ordinary testing purposes, to run more than 10 million cycles. If the specimen does not break within 10 million, it is at, or very close to, a stress at which it will run to many billions or indefinitely, in the absence of corrosion. However, some non-ferrous alloys, notably Monel and duralumin, annealed and cold-drawn copper, and annealed 60:40 brass may show no clear endurance limit on logarithmic or semi-logarithmic plotting of the S-N curves, though the tests are extended to 100 million, or even 500 million cycles, though some non-ferrous alloys may show endurance limits at less than 20 million cycles.[16, 20]

Memmler[21] says that, on steel 10 million, on nickel 100 million cycles, and on other non-ferrous metals perhaps even more, are required. Judge[22] says that 50 million are required for non-ferrous alloys. On copper-beryllium alloys, the N. P. L.[23] found that specimens fractured after running as many as 65 million cycles. It would appear that tests should be carried to at least 100 million cycles, or to at least 50 million with subsequent

TABLE 1B. EFFECTS OF P, As, Sb IN ANNEALED COPPER

							STATIC				ENDURANCE	
No.	Worker	Mark	Composition			Anneal	Tensile Strength lbs./in.^{2}	Elong. %	Red. Area %	Impact ft.-lbs. Izod (Bars not Broken)	Stress lbs./in.^{2} for 20 Million Cycles	Endurance Ratio
	Effect of Phosphorus											
			% O	*% P*								
1	Hanson, etc.	CP14	.019	.014		1290°F. ½ hr.	34,500	32	72½	39	11,000	.32
2		CP11	.010	.03		1290°F. ½ hr.	32,000	58½	81½	37½	12,000	.38
3		CP3	.009	.045		1290°F. ½ hr.	32,500	50	85½	41	12,500	.39
4		CP4		.096		1290°F. ½ hr.	33,000	61½	79½	40	14,250	.43
5		CP5		.148		1290°F. ½ hr.	34,000	63	85	41	15,000	.44
6		CP6		.178		1290°F. ½ hr.	35,000	60½	85	43	13,250[a]	.38
7		CP7		.254		1290°F. ½ hr.	35,500	62½	84	40½	13,500[a]	.38
8		CP8		.494		1290°F. ½ hr.	38,500	62	89½	42½	15,500	.40
9		CP22		.69		1290°F. ½ hr.	38,500	62½	83½	44	16,500	.43
10		CP23	.002	.79		1290°F. ½ hr.	40,000	64	81	44	17,500	.44
11		CP24		.95		1290°F. ½ hr.	40,000	65½	84½	44	17,000	.42
	Effect of Arsenic Plus Oxygen											
			% O	*% As*								
12	Hanson and Marryat[14]	20	.11	.09		1290°F. ½ hr.			58	39	15,500	—
13		22	.039	.09		1290°F. ½ hr.	32,500	62	70	43½	13,500	.42
14		23	.04	.24		1290°F. ½ hr.	33,000	57	71	43	15,500	.47
15		8	.06	.25		1290°F. ½ hr.	32,500	57½	66½	40	15,000	.46
16		7	.07	.30		1290°F. ½ hr.	33,000	55	65½	38½	15,000	.45
17		19	.058	.34		1290°F. ½ hr.	33,500	55½	64	38	14,500	.44
18		6	.071	.44		1290°F. ½ hr.	33,000	58½	67	38½	15,500	.47
19		24	.034	.45		1290°F. ½ hr.	34,000	62	71½	42	16,000	.45
20		25	.05	.93		1290°F. ½ hr.	35,500	61	71½	44	15,500	.44
21		39	.0052	1.40		1290°F. ½ hr.			67	40	15,500	—
22		21	.006	2.02		1290°F. ½ hr.	36,500	58½	64	42	17,000	.47
	Effect of Antimony											
			% O	*% Sb*								
23	Archbutt & Prytherch	CS2	.008	.0035		1290°F. ½ hr.	31,500	62½	74½	40	12,000	.38
24		CS3	.013	.021		1290°F. ½ hr.	32,000	62½	73½	37	13,000	.40
25		CS4	.005	.046		1290°F. ½ hr.	32,000	60	72	39	13,000	.40
26		CS5	.015	.092		1290°F. ½ hr.	33,500	48½	72½	40	13,000	.39
27		CST3	.016	.22		1290°F. ½ hr.	33,000	67	77	35	15,500	.47
28		CST4	.014	.47		1290°F. ½ hr.	33,500	57½	66	36	17,500	.52
	Effect of Arsenic Plus Antimony											
			% O	*% As*	*% Sb*							
29		CAST1	.014	.052	.050	1290°F. ½ hr.	32,000	60	73	37	15,000	.47
30		CAST9	.018	.045	.32	1290°F. ½ hr.	32,000	58½	71	36½	15,000	.47
31		CAST3	.019	.048	.51	1290°F. ½ hr.	33,000	57½	73	41	16,000	.49
32		CAST6	.018	.249	.05	1290°F. ½ hr.	34,000	59	76½	37½	16,000	.47
33		CHST4	.017	.244	.32	1290°F. ½ hr.	33,500	59	75	41	16,000	.48
34		CHST5	.013	.24	.55	1290°F. ½ hr.	34,000	64	76	33½	17,500	.51
35		CAST10	.018	.49	.06	1290°F. ½ hr.	35,000	60	75½	12½[b]	17,000	.49
36		CAST8	.017	.50	.53	1290°F. ½ hr.	36,000	62	75½	15½[c]	17,500	.49

[a] Specimens CP6 and CP7 not entirely sound

[b] Low value—repeat test gave 37½.

Low value—repeat test gave 34½.

raising of the stress to see whether strengthening by under-stressing or damage by over-stressing has occurred.

It may be true in any given case that 20 million cycles may give enough information for purposes of routine testing, but one cannot say so before hand without having made longer runs, or unless a clearly marked "knee" in the S-N curve has been reached. The S-N curves for non-ferrous alloys are generally flatter than those for any but the softest ferrous materials. There is less tendency to show a "knee" in the curve, i. e., a fairly sharp change in slope with the curve beyond the "knee" becoming recognizably asymptotic to the stress corresponding to the endurance limit; hence it is harder to determine whether a stress has been determined that, with the application of a suitable factor of safety, is useful for purposes of design.

Dr. McAdam believes that the presence or absence of a knee means little, feeling that the shape of the curve above the endurance limit is a function of the form of the test specimen as well as of the material itself, and evaluates the presence or absence of a definite endurance limit on the basis of the approach of the curve to the stress asymptote. He believes that, in general, 100 million cycles will serve to establish the endurance limit for non-ferrous alloys as well as 10 million cycles does for steel. The R. R. Moore specimen is stated by McAdam to have more tendency to show a kneed curve than the tapered cantilever specimen generally used by McAdam.

Few sufficiently long-continued tests have been made on many copper-base alloys, and while it would be premature, in the writer's opinion, to assume that the 100 million cycle figures represent absolutely definite endurance limits, in general such figures are probably adequate for engineering purposes on most copper-base alloys. There are cases where the stress for a life of 20 million cycles is little if any higher than that for 100 millions, but these cases are rather rare, and most of the endurance limits given in the literature for 20 million cycle tests should be shaded downwards, while those for 10 million or less can be used only as rough indications and for comparison with the high-stress end of real endurance curves.

Since the endurance ratio for non-ferrous materials is not constant one cannot even roughly predict from the static properties what the endurance will be. While material of higher ductility may show greater life on the high stress end of the S-N curve, there is no clear indication in the data so far found in the literature that the actual endurance limit rises in proportion to increase in ductility. Nor, after the endurance limit for a sufficient number of cycles has been determined on a properly filleted and polished specimen, can one exactly predict what factor of safety must be allowed over the endurance figure if the material is actually to be used with poor fillets, surface notches, or other stress-raisers, for adequate data are lacking as to the relative susceptibility of different alloys in different tempers to the "notch effect." It is known that tough pitch copper is quite susceptible to it, some of its alloys somewhat less so, but in general, our ignorance on this subject is abysmal.

Moore[24] states that very pure metals and very fine-grained alloys are more sensitive to the effect of notches, and brings out the fact that mere ductility does not necessarily give freedom from notch-damage, but that a property he calls "crackless plasticity" is the essential. This is in some way related to the "damping capacity" of a metal. Damping has been discussed by von Heydekampf[25] and studied in great detail by the Wöhler Institut in Germany. So little has been done in the correlation of damping tests with the endurance limit and the susceptibility to notch-tenderness that about all that is certain is that damping tests, as clearly pointed out by von Heydekampf, do not serve as short-time endurance tests but hold chief promise for throwing light on the notch effect. Few of the Wöhler Institut publications on damping give the endurance properties of the alloys studied. One of these, by Schmidt[26] gives a few such comparisons, but the actual endurance tests are carried only to 2 million cycles, so that the data give little information on true endurance properties. One by Föppl and Schaaf[27] deals with damping of copper and of a brass whose composition is not given, but gives no endurance data.

With the probable exception of the high-speed fatigue method of the Bureau of Standards, which appears applicable to some of the copper-base alloys in the form of sheet, there seems to be no sufficiently acceptable short-cut method, and since the tests should be carried to say 100 million cycles, it is no wonder that information is still scanty. One cannot necessarily assume that results on rod are applicable to sheet of the same static properties; so testing of different forms of the same alloy is required. Most endurance data for non-ferrous alloys are fairly modern, and so the data available for rotating beam specimens are presumably for specimens with the accepted type of fillets and finish, run at normal speed. Data on the rotating cantilever (McAdam), on the Farmer-Sondericker-R.R. Moore type and the Illinois spring machine type are considered equally acceptable. These are constant-stress methods. While the constant-strain method used by Townsend[28] is the one most easily applied to thin material, and the uses of the material in service may make the test more analagous to service than a constant-stress method would be, the results are not expressible in lbs./in.^{2} with quite the same degree of certainty, since the modulus of elasticity must be accurately known, and must not change during the run, for the results to be accurate.

Summary. The additions to copper that are of known value in increasing the endurance properties consist of phosphorus, arsenic, and antimony, all of which lower the electrical conductivity.

No data are available on oxygen-free copper. Within the tough-pitch range the percentage of oxygen appears to have

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TABLE 2. APPARENT EFFECT OF OXYGEN, ARSENIC, AND IRON ON ANNEALED COPPER BY A NON-STANDARD, SHORT-CUT METHOD OF QUESTIONABLE VALIDITY

No.	Worker	Mark	Composition			Anneal	STATIC Tensile Strength lbs./in.^{2}	STATIC Elong. %	STATIC Red. Area %	STATIC Impact ft.-lbs. Izod (Bars Not Broken Unless Noted)	Assumed Endurance by Short Cut Method (Not True Endurance Test) lbs./in.^{2}
	Effect of Oxygen		*% O*								
1	Gough[4]		.04			Annealed	29,000	60			12,500
2			.06			Annealed	29,250	56			12,500
3			.09			Annealed	29,500	52½			12,500
4	Hanson, etc.[5]	C35	.016			1290°F. ½ hr.	32,500	53½	77	46	11,000
5	These specimens were made in the laboratory and were not entirely sound	C28	.04			1290°F. ½ hr.	32,000	50	71½	45½	13,500
6		C30	.06			1290°F. ½ hr.	32,500	55	70	47½	13,000
7		C27	.09			1290°F. ½ hr.	33,000	52½	64½	45	12,000
8		C25	.17			1290°F. ½ hr.	34,500	49	56½	31	11,000
9		C36	.36			1290°F. ½ hr.	37,000	34½	38½	16 (broke)	11,000
	Effect of Arsenic		*% O*	*% As*	*% Fe*						
10	Hanson and Maryatt[5]	AC8	.016	.053	.008	1290°F. ½ hr.	31,500	56½	72	46	14,000
11		AC7	.005	.093	tr.	1290°F. ½ hr.	32,000	56½	70	43	14,500
12		AC9	.003	.36	tr.	1290°F. ½ hr.	32,500	60	78½	46	13,500
13		AC10	.009	.60	tr.	1290°F. ½ hr.	33,500	55	61½	17½ (broke)	14,500
14		AC5	.013	.86	tr.	1290°F. ½ hr.	34,000	56	65½	30	15,000
15		AC4	.006	1.04	tr.	1290°F. ½ hr.	34,000	59	78½	47	15,500
	Effect of Iron		*% O*	*% Fe*							
16	Hanson and Ford	F01	.014	0.06		1290°F. ½ hr.	32,500	57	73	44	14,000
17		F04	.003	0.2		1290°F. ½ hr.	32,000	60	73	43	13,500
18		F06	.004	0.4		1290°F. ½ hr.	33,500	60	80	43	14,500
19		F08	.008	0.73		1290°F. ½ hr.	37,500	52	80	51	14,500
20		F1	.005	0.96		1290°F. ½ hr.	36,000	45	82	63	15,000
21		F15	.004	1.38		1290°F. ½ hr.	43,000	29½	79	58½	15,500
22		F2	.007	1.80		1290°F. ½ hr.	44,500	29	79	65	16,000
23		F25	.008	2.09		1290°F. ½ hr.	49,500	33½	79	70	20,000